TRANSACTIONS
American Society for Steel Treating

VOL. X NOVEMBER, 1926 NO. 5

ORDNANCE PROGRESS

ON the 8th day of October, 1926, the Army Ordnance Association held its eighth annual meeting at the Aberdeen Proving Ground of the United States Army Ordnance Department, Aberdeen, Maryland.

Several other organizations were invited by the Army Ordnance Association to attend this meeting. These were the National Association of Manufacturers, the National Industrial Conference Board, The Reserve Officers' Association of the Unites States and the American Society for Steel Treating. Approximately 2000 people attended this meeting which consisted of an all day program of demonstrating the state of preparedness of the United States army. It was a well rounded out program, starting at 10 a. m. and continuing until 8 p. m. The various departments of the army participating in this meeting and demonstrating the various equipment were the Ordnance Department, Chemical Warfare Service, Field Artillery, Coast Artillery, Army Air Corps, Signal Corps and Corps of Engineers, United States Army.

A main feature of the many demonstrations in the use of arms and equipment was the first on the program, that of firing one round from a 16-inch gun mounted on a barbette carriage. This gun is a weapon recently constructed for seacoast defense in this country and has a maximum range with a charge of 860 pounds of powder of over 30 miles and fires a projectile weighing 2100 pounds at a speed of 2750 feet per second. It is capable of sustaining a rate of fire of one round per minute. This gun is an armor piercing weapon capable of piercing 16 inches of armor plate and having a flat trajectory. It is mounted in the open without parapets. The gun has electrical elevating and traversing devices and a mechanical rammer. It has been adopted as the standard seacoast defense gun and is being installed at the principle harbor defenses of the country.

The second interesting demonstration was that of the firing of a 12-inch gun on a railway mount used as coast defense mobile artillery. This type of carriage was designed for land warfare using a 12-inch 35-caliber gun firing a 700-pound projectile. The emplacement consists of a ground platform, two of which are carried with the mount. Upon this emplacement the mount has 10 degrees of traverse and can fire at angles up to 38 degrees in elevation with the corresponding range of 16 miles for the 700-pound projectile.

A most interesting demonstration during the morning was that of antiaircraft firing by the 61st Coast Artillery (Antiaircraft Regiment) from Fort Monroe against sleeve targets using the latest types of antiaircraft matériel.

The firing was done with a battery of four 3-inch antiaircraft guns of latest type. These firings were made against a sleeve target towed by an airplane. The guns were mounted in a fixed emplacement and were the approved type for seacoast and fixed defenses. These guns fire shell or shrapnel at 2800 foot-seconds.

The next event was a demonstration of firing a 37-millimeter antiaircraft gun at a towed target. The gun was a Browning automatic 37-millimeter, model 1925, designed for 3000 foot-seconds velocity. The gun was mounted on a 37-millimeter antiaircraft mount which permits all around fire and an elevation of 90 degrees.

Following this demonstration firings were made from .30 and .50 caliber machine guns at towed targets. The target was about 14 feet long, 3 feet in diameter, towed on a cable 1800 feet long and moving at approximately 100 miles an hour.

Tracer bullets were used in this firing in order that the line of fire could be easily seen. It was indeed interesting to observe the accuracy with which these bullets were directed at the target, the target being literally riddled.

After these demonstrations a few rounds were fired from a 105-millimeter antiaircraft gun. This is a pilot mount recently received at Aberdeen Proving Ground for test. It fires a 33-pound projectile at a vertical range of 7 miles and a horizontal range of 11 miles.

After this demonstration of antiaircraft firing the audience moved to the aviation field where they witnessed a very interesting demonstration of parachute jumping as used in landing from a

moving plane and the laying of a smoke screen with an airplane.

Then moving to the main front the crowd witnessed the dropping of 600-pound demolition bombs from a Martin bomber. These bombs are intended for use in attacking ammunition and supply dumps, storehouses, terminals and similar targets. The bombs are equipped with instantaneous fuses for land impact.

The method of dropping smoke bombs from an airplane for the purpose of forming a smoke screen was also demonstrated.

After luncheon the audience moved to the main front where they witnessed a most interesting demonstration of automotive equipment. A brief description of each of these pieces was made by the announcer. The vehicles were then put into operation maneuvering over a small area of ground in clear view of the spectators.

Tanks, tractors for gun carriages, self-propelled guns and cross country passenger and cargo vehicles gave demonstrations separately and simultaneously.

One of the interesting demonstrations of this part of the program was the bombardment and destruction of a stronghold defended with machine guns by a large tank of the type developed by the joint British and American commission and used against the trench system of the Hindenburg line.

Another interesting feature was the 75-millimeter gun motor carriage designed to mount either a 75-millimeter gun or a 105-millimeter Howitzer. This carriage has a speed of 30 miles per hour on good roads.

After this demonstration a battalion of the 6th Field Artillery maneuvered into position and registered on a target with shrapnel and then put down a smoke barrage of 16 rounds from each field piece.

Following this a demonstration of ammunition was made showing the firing of a 37-millimeter gun using a supersensitive fuse against airplane fabric. These shells were so sensitive that the target was demolished after 3 or 4 shots.

In order that the marked contrasts between the old type of powders and the present be shown, a demonstration was made by firing shots with black powder and then with smokeless powder. Many other types of firings were then made illustrating the numerous sizes of guns and the method of determining the character-

istics of the shells in flight, resistance due to the air and its effect upon the flight of projectiles. The various types and sizes of trench mortars were fired using smoke shells, so that the audience could readily see where the shells landed.

After this demonstration, the audience adjourned to the museum building wherein are kept many interesting examples of army equipment some of which were captured during the World War. At 4 o'clock the Army Ordnance Association held its annual business meeting.

At 6:15 the night firing maneuvers were started. The first feature of this part of the program was the firing of a 75-millimeter gun with one round of black powder, one round of smokeless powder and one round of flashless powder. This same demonstration was repeated with a 155-millimeter gun.

Following this a demonstration of antiaircraft night firing by the 61st Coast Artillery (A. A. Regiment) took place using the latest type of matériel. These firings were made on sleeve targets towed by airplanes and located by means of sound locators and illuminated by a battery of the latest type of antiaircraft searchlights. These searchlights were 60-inch Sperry type illuminated with 800,000,000 candle power throwing an effective beam of light 7 miles in length. By means of the Vickers range finding instrument and sound locators the airplane was located and the searchlights trained upon the target. The locators are connected with the searchlights with a comparator by which the light is directed to the position of the target. Then by means of the Vickers range finder the range and speed of the target is transmitted to the dials on the guns. This instrument is capable of controlling the range of 100 guns at one time. The first firings were from a battery of four 3-inch guns in which about 60 shots were fired at two flights of the target. Following this a battery of eight .30-caliber machine guns fired 1600 shots using tracer ammunition in the ratio of 1 to 12. This demonstration was most spectacular.

The Army Ordnance Association is to be congratulated for a most successful and interesting meeting and those who had the privilege of seeing the state of preparedness of the various departments of the army and the way in which modern warfare is conducted, came away with a certain sense of safety and security as far as modern defense is concerned.

REPORT OF SUB-COMMITTEE ON MEASURING CASE DEPTH

THE Recommended Practice Committee authorized the appointment of a sub-committee to compile a recommended practice for the measuring of case depth. The sub-committee has been actively engaged with this problem for some time and recently submitted to the Recommended Practice Committee a tentative report.

It was the committee's endeavor to definitely recommend a standard method for measuring case depth, but upon careful study of the various methods used, the committee felt that it was inadvisable to make definite recommendations, because of the conflicting opinions of just what constitutes the exact depth of case, and because the depth of case of a carburized object may vary depending upon the methods used; the compounds used and the heat treatment of the metal at the time of measuring the case.

The following report was compiled to serve as a basis for a standard recommended practice. The report contains the various methods used but no attempt has been made to make a definite recommendation. In order that the sub-committee may establish a recommended practice that outlines explicitly a convenient and accurate method for measuring case depth, comments and suggestions on the following report are solicited.

Comments should be mailed to the secretary of the committee, J. E. Donnellan, 4600 Prospect Ave., Cleveland, Ohio.

The personnel of the sub-committee is as follows:

S. P. Rockwell, Chairman
F. C. Raab
B. F. Shepherd
R. W. Woodward
R. J. Allen

SUGGESTIONS FOR MEASURING CASE DEPTH

There are many methods of measuring the depth of penetration of carbon of carburized and case-hardened steels. There is no standard method used. Definitions made by a few members of this or any other society are probably not acceptable to all, but it is hoped that these remarks will serve as a basis for discussion and final acceptance of satisfactory definitions.

A case depth may vary considerably depending on the method

used, the state of heat treatment of the metal at time of measurement, and the physical requirements desired of the case. A number of the various and common methods used are given as follows:

1. Scale measurement of ordinary fracture.
2. Scale measurement of blued fracture. The bluing is secured by heating the fractured specimen slowly until the temper colors appear. The higher carbon sections of the case reach a blue color before the lower carbon sections. Bluing is best done on an electric hot plate.
3. Scale measurement of ordinary fracture which has been dipped in acid. Nitric acid is usually used, either strong or diluted with water or alcohol. The higher carbon sections of the case discolor before the low carbon sections.
4. Scale measurement of polished cross-sections.
   (a) Blued as above.
   (b) Etched in acid as above.
5. Microscopic measurement of highly polished cross-sections etched in standard reagents.
6. Filing or drilling through from the core at right angles to the case until the case hardness resists further cutting or abrasion. Measurement of the unfilable case remaining is taken as the depth. This method, of course, depends on the removal, say by grinding, of enough of the case to be measured, so that a drill or file may be used.
7. Machining off successive amounts of surface from softened carburized work, analyzing and basing results of depth on individual conception of the amount of carbon which can be considered case.
8. Weighing the parts to be carburized before and after carburizing, and computing the added carbon and depth of penetration.
9. Magnetic methods.
10. Taper grinding work hardened so that some unit of lineal measurement of surface represents a known depth from original surface. Testing each unit of lineal surface by some practical form of hardness test. This not only shows depth, but characteristics of the different depths from the surface in terms which are readily convertible to resistance to wear and deformation.

Discussion of Methods

It must first be recognized that the total case is, strictly speaking, composed of all added carbon over the core content, and decreases from a maximum of carbon at the surface to the carbon of the core. For this reason it must be expected that the total case will not all harden to the same degree and that the effective case is that which receives maximum hardening, is reasonably uniform in hardness, and is useful in fulfilling the purpose for which the article has been case-hardened.

Most of the above methods determine the depth as some value between the effective case depth and the total case depth.

Methods 1, 2 and 3, which measure depth by reference to fracture, are subject to considerable variation depending upon the heat treatment and subsequent carbon diffusion, and by the personal equation of the operator in conducting the test. The usual run of commercial steels give variations in fracture tests due to the allowable carbon range of the core. Fractures showing crystalline cores are harder to estimate than those having amorphous or fibrous cores.

Method 4 is subject to similar errors as 1, 2 and 3, but eliminates some of the errors possible in the latter methods.

In Method 5 the depth may vary due to the observer's ideas as regards what percentage of carbon makes up the case, and how much increase in carbon is allowable for core. The division between the eutectic zone and the hypoeutectic is readily discerned and probably bears a close relation to the "effective case" depth.

Method 6 depends upon the use of a file and is subject to the personal equation attendant thereon.

The method of chemical analysis, No. 7, is quite costly and is subject to the same limitations as Method 5 using the microscope.

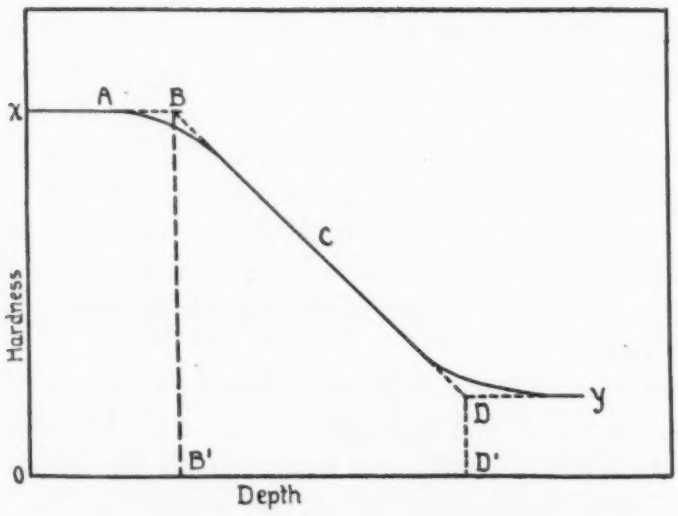
Weighing (Method 8) has the advantage of being nondestructive but involves complicated calculations.

Magnetic methods (9) are also nondestructive, but it is felt that more research is necessary before these methods are available for commercial routine.

The committee feels that Method 10 offers the greatest possibility of furnishing a standard procedure as it is not subject to personal errors and fulfilling in a large part the accepted purposes of case-hardening as regards total depth of case and "effective

depth''. This method can be applied to a finished article and any control method adopted may be referred to it by the use of a suitable constant. This constant will depend upon the variables mentioned above, and particularly upon the place in the heat-treating operations at which the control method is applied.

Fig. 1 shows a typical hardness-depth curve obtained by the suggested standard method (No. 10). The curve XCY may be divided by tangents into three distinctive parts, point A being the maximum hardness obtained and AB a tangent at this point. DY



is the tangent to the hardness of the core, and BCD the tangent to the intermediate hardness zone. The intersections of these tangents and their abscissae OB′ and OD′ represent respectively the ''effective case depth'' and the ''total case depth''. This conception places the determination of case depth upon a purely mathematical basis and takes into account the hardness of the article as it will be put into service.

This method is not a control method to be used in the hardening room, but is a suggestion for a standard method to which the control method may be referred with the proper constant.

As a satisfactory control method which can be applied at the time the carburizing process is being carried out, the committee recommends that a sample be withdrawn from the carburizing heat, water quenched, fractured and the depth determined by Method 1. Due to subsequent heats which may diffuse the carbon, lower hardening temperatures, a different quenching medium, and

grinding operations, the depth determined by the control method may not coincide with the depth determined by a standard method on the finished article, and the predetermined constant or factor must be applied at this time.

WINTER SECTIONAL MEETING

THE Winter Sectional Meeting of the American Society for Steel Treating will be held in Washington, D. C., January 20 and 21, 1927. This meeting will be under the auspices of the Washington-Baltimore Chapter of the society.

Plans and arrangements for this meeting are well under way and judging by the interest and enthusiasm which has been shown by the members of the Washington-Baltimore Chapter and especially their executive committee and sectional meeting committee this will be a most successful and interesting meeting. A comprehensive technical program and an interesting plant inspection program has been arranged. These will be announced in the next issue of TRANSACTIONS. Headquarters will be at the new Mayflower Hotel.

The committee on arrangements is as follows:

General Chairman—Jerome Strauss
Finance—L. H. Fawcett, G. W. Quick
Plant Inspection—J. E. Crown, J. R. Freeman, Jr. Emil Gathmann
Meetings and Papers—H. J. French, R. T. Bayless
Hotel—H. K. Herschman, H. C. Cross
Entertainment—W. L. Blankenship
Local Attendance—Samuel Epstein
Publicity—H. W. Gillett

Future Meetings

WINTER SECTIONAL MEETING—Washington, D. C., January, 20 and 21, 1927.
SPRING SECTIONAL MEETING—Milwaukee, May, 1927.
ANNUAL CONVENTION AND EXPOSITION—Convention Hall, Detroit, September 19-23, 1927.

UNAUDITED PROFIT AND LOSS STATEMENT AMERICAN SOCIETY *for* STEEL TREATING

For the period from January 1 to September 30, 1926

INCOME			
Gross Dues		$35,633.43	
Transactions Advertising	$39,872.38		
Transactions Sales	891.46		
Transactions Subscriptions	982.82	41,746.66	
Reprints		633.65	
Bindery Account		823.11	
Book Account		1,540.66	
Pencil Account		36.25	
Pins and Buttons		37.00	
Miscellaneous Receipts		222.22	
Data Sheets	1,219.07		
Data Sheet Binders	916.15	2,135.22	
Discounts Received	508.32		
Interest	1,647.93		
H. M. Howe Medal Fund Income	61.98	2,218.23	
Sustaining Exhibit Membership		8,375.00	
TOTAL INCOME			$93,401.43
EXPENSE			
Local Chapters		$17,319.92	
Transactions		26,881.73	
Reprints		1,012.15	
Bindery Account		850.02	
Book Account Expense		1,233.59	
Pencil Account		64.04	
Library		117.32	
Data Sheets		3,611.05	
Data Sheet Binders		583.20	
Discounts Allowed		360.84	
Bank Exchange		3.68	
E. D. Campbell Memorial Lecture		625.60	
General Expense		4,926.54	
Secretary's Office		7,898.39	
President's Office		1,643.21	
Treasurer's Office		2,436.39	
Directors Expense		2,032.13	
National Committees		2,954.58	
Old Accounts Cancelled		115.00	
TOTAL EXPENSE			$74,669.38
EXCESS INCOME OVER EXPENSE			18,732.05

BALANCE SHEET

As of September 30, 1926

ASSETS			
Commercial Account—Cleveland Trust Co.		$ 8,142.37	
Savings Acounts—Cleveland Trust Co.	$15,408.12		
Equity Savings & Loan Co.	17,966.14		
Union Trust Co.	5,195.04	38,569.30	
Accounts Receivable—Advertising	12,481.74		
Miscellaneous	11,047.16	23,528.90	
Investments		63,235.00	
Office Furniture and Fixtures		798.30	
Inventory		2,452.65	
1926 Convention Prepaid Expenses		56,571.01	
1927 Convention Prepaid Expenses		482.62	
TOTAL ASSETS			$193,780.15
LIABILITIES, RESERVES AND SURPLUS			
Accounts Payable		1,476.84	
1926 Convention Advance Receipts		87,204.11	
Reserves For:			
Past Due and Doubtful Accounts	890.25		
Dues Paid in Advance	10,000.00		
Permanent Convention Reserve	20,000.00		
H. M. Howe Medal Fund	5,000.00	35,890.25	
Surplus—			
January 1, 1926	52,476.90		
Less Increase to H. M. Howe Medal Fund	2,000.00		
	$50,476.90		
Plus—Excess Income for period from January 1 to September 30, 1926, per Profit and Loss statement attached	18,732.05		
Present Surplus		69,208.95	
TOTAL LIABILITIES, RESERVES AND SURPLUS			$193,780.15

WEAR OF STEELS WITH PARTICULAR REFERENCE TO PLUG GAGES

By H. J. French and H. K. Herschman

Prefatory Abstract

A new gage wear testing machine is described and results obtained by its use are discussed for wet sliding friction of metal on metal and wear by abrasives. Comparisons of service wear tests at different plants with results obtained in the laboratory with the gage wear tester are included. Data are presented to show that file hardness is not a criterion of good wear resistance of gages. Chromium-plated steel gages are likewise compared with gages made from heat treated steels of the customary types.

I. Introduction

TESTS described in this paper represent part of an investigation of steels and treatments best suited for gages. The request for a study of gage steels by the Bureau of Standards was made through Col. J. O. Johnson, Chief of the Gage Section, Ordnance Department, U. S. A., simultaneously with the formation in 1922 of a group known as the Gage Steel Committee, consisting of representatives of gage producers and consumers, steel manufacturers and government departments.

The problem presented was divided into three parts as follows: (1) control and possible minimization of dimensional changes in hardening, (2) to secure dimensional permanence, i. e., the minimum of dimensional changes with time, and (3) a study of steels and treatments giving the greatest resistance to wear.

The first two phases were studied in considerable detail by Howard Scott under the general guidance of the Gage Steel Committee and in addition to periodic progress reports a number of publications were prepared[1]. The study of gage wear until re-

[1]Howard Scott: Quenching Properties of Glycerine and Its Water Solutions. Transactions, American Society for Steel Treating, 6, p. 13, 1924. Dimensional Changes Accompanying the Phenomena of Tempering and Aging Tool Steels, Transactions, American Society for Steel Treating, 9, p. 277, 1926. Origin of Quenching Cracks. Bureau of Standards Scientific Papers, No. 513, September, 1925.

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Of the authors, H. J. French is senior metallurgist and H. K. Herschman, assistant scientist, Bureau of Standards, Washington, D. C.

cently received the least attention, and it is this phase of the general problem which is considered in this report. However, the tests to be described are only of a preliminary nature, and are to be followed by further work with the gage wear tester recently devised by the authors, as well as by additional service tests.

II. Previous Studies of Wear

1. *Methods of test*

Among the recognized causes of wear are the following:

A. Rolling friction
   1. lubricated—as, for example, that between steel balls and the race in ball bearings
   2. unlubricated—such, for example, as that between a carwheel and rail.

B. Sliding friction
   1. lubricated—for example, that between a shaft and plain bearing
   2. unlubricated
      a. between 2 solid bodies such as a brakeshoe and wheel or a wheel and rail during braking
      b. between a solid body and an abrasive material of more or less fine grain such as that to which grinding, crushing and excavating machinery is subjected.

Aside from records of performance in actual service, a variety of laboratory wear testing machines have been devised for the study of the different types of wear, and some of these are shown in Fig. 1. The diagrammatic sketches given do not contain details of construction, but are intended merely to illustrate in a general way the methods by which different types of wear have been produced in the laboratory by various investigators.

The machines used by Robin (21)[2], Brinell (7), Honda and Yamada (4) and the ball mill, all illustrated in Fig. 1, give so-called "abrasive wear", i. e., wear of type B2b in the classification above.

The Derihon "abrasion mill" (19) and the Jannin equipment (12) (13) give wear of class B1, by sliding lubricated friction.

Wear by unlubricated rolling friction is obtained by the ma-

[2]The numbered references refer to the selected bibliography appended to this report.

chines of Saniter (18), Stanton and Batson (15) and Amsler (5) (9) (11).

The different machines shown graphically in Fig. 1 have been classified according to their original applications, but by modification in details of construction they can, in some cases, be used

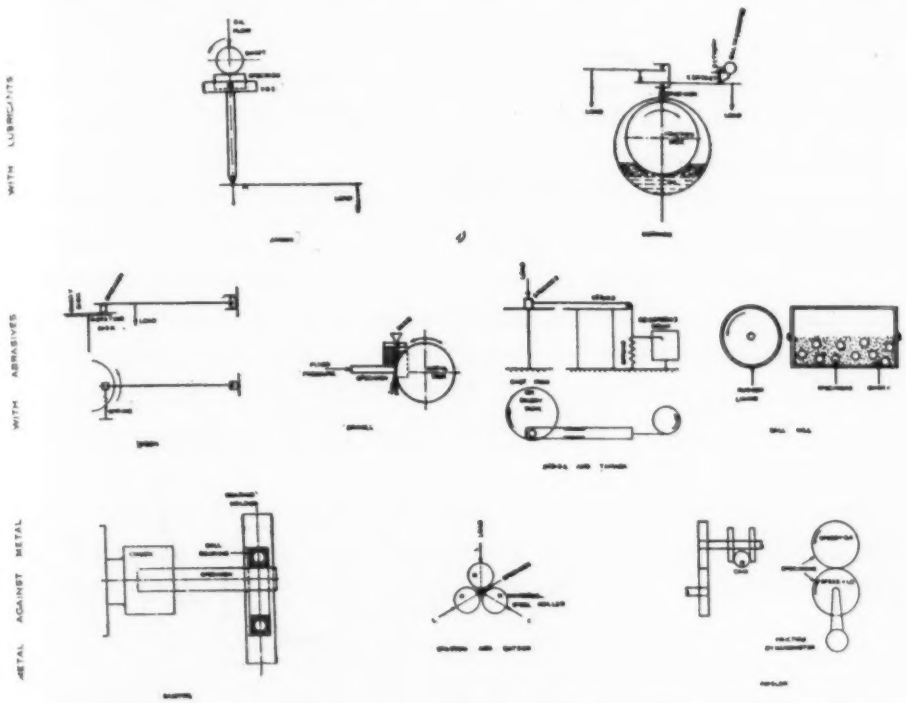


Fig. 1—Sketches Illustrating Laboratory Wear Testing Machines Used by Various Investigators.
Detailed Descriptions of these Machines as well as Results Obtained by Their Use are Given in Reports Included in the Appended Bibliography.

to produce wear of different types. For example, the Amsler machine can be used with a lubricant and it is likewise possible to employ the Jannin method of construction for a study of wear by unlubricated sliding friction. Whether or not such modifications would be desirable is open to question, but the possibility of extending the application of the different test units remains.

Insofar as steels are concerned, most attention has been given to abrasive wear (Robin, Brinell, Honda and Yamada and ball mill tests) and to unlubricated rolling friction (Saniter, Stanton and Batson and tests with the Amsler machine). The study of wear by lubricated friction has so far been restricted largely to bearing metals because, industrially, the bronzes are ordinarily in contact with "harder" steels and under such conditions wear faster.

The wide variation in the nature of the machines used by different investigators for the study of wear of metals is of itself evidence that different effects are to be expected under different conditions, and this is readily confirmed by a study of the results

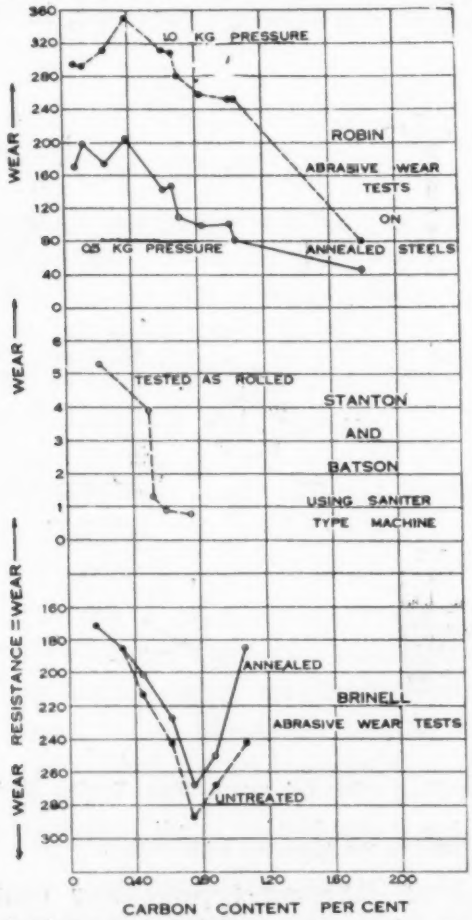


Fig. 2—Effect of Carbon Content on the Wear of Carbon Steels as Determined by Various Investigators.
The Methods of Test Employed by the Different Investigators are Shown in Fig. 1.

obtained with the different machines. As an example there is given in Fig. 2 a summary of the effect of carbon content on the wear of carbon steels in the annealed or "natural" conditions, as determined by Brinell and Robin in abrasive wear tests and by Stanton and Batson with the Saniter test method (unlubricated rolling friction). There is likewise given in Table I a summary of the effect on wear resistance of small additions of special elements to steels in the annealed or "natural" conditions. These comparisons show clearly that great care must be used in the in-

terpretation of test data and that it is not yet safe to attempt predictions from one set of conditions to another.

2. *Indentation hardness versus wear*

Despite differences in the effects observed by different investigators with various test methods, there are a number of general conclusions which may be drawn from the early studies of wear. Resistance to wear does not, in general, increase directly with tensile strength, indentation hardness (Brinell method) or hardness determined by other of the customary methods. Both indentation hardness and toughness are important factors affecting wear. High indentation hardness tends to decrease wear as it increases resistance to penetration, scratching or displacement of the metal. On the other hand, toughness tends to decrease wear by making it difficult to tear off from the metal the small particles which may be displaced. Wear usually takes place by combined displacement and removal from the main body of the metal of such small particles.

The relative importance of indentation hardness and toughness will be expected to vary depending upon the conditions under which wear takes place and the character of the metals involved.

3. *Metallurgical factors affecting wear*

According to Robin (21) the "metallurgical origin of the metal" or "the degree of refining which the material has under-

Table I

Effect of Small Additions of Special Elements, Such as Are Ordinarily Found in Commercial Structural Steels, on Wear Resistance as Determined by Different Investigators.[1]

Variable	How wear resistance is affected by the variable in column 1.			
	Abrasive Wear Tests		Unlubricated Rolling Friction	
	Robin	Brinell	Saniter	Norris
Small additions of				
Manganese	reduced	improved	little influence	improved
Silicon		reduced		
Nickel	reduced	improved	improved	improved
Chromium	improved	improved		improved
Tungsten	improved			
Vanadium	improved			improved

[1]This applies only to steels in a "soft" condition, i. e., as hot finished, tempered at high temperatures subsequent to hardening, annealed or normalized. For sources of information, refer to the bibliography appended to this report.

gone can be revealed by abrasive tests. Crucible steels and steel from electric furnaces are less susceptible to wear than open-hearth steels. Basic steel appears to offer the lowest resistance to wear—". On the other hand, Brinell (7), in a set of abrasive wear tests of acid bessemer and acid open-hearth steels made in one plant, found that "differences in process of manufacture" could not be "considered as contributory causes" of differences in wear.

Differences in wear resistance have been ascribed to variations in grain size and the presence of impurities such as slag. For example, in abrasive wear tests of two lots of Lancashire iron Brinell (7) found that "the wear resistance investigated perpendicular to the rolling direction was higher than in the rolling direction."

Robin (21) found that "the degree of resistance to wear is increased by working (hammering, rolling, etc.)", and likewise that fine-grained martensites frequently showed better wear resistance than coarse-grained martensitic structures. Brinell (7) observed that increase in the annealing temperature between about 1470 and 2190 degrees Fahr. (800 and 1200 degrees Cent.), which presumably increased grain size, improved wear resistance but in all such cases the Brinell hardness was also found to be higher. Where no appreciably higher Brinell numbers were obtained from increase in the annealing temperature, no appreciable difference was observed in the wear by abrasives.

Since both indentation hardness and toughness affect the wear of metals, factors such as chemical composition and heat treatment which affect their indentation hardness and toughness may be expected to affect their wear resistance. These features are emphasized in many of the investigations referred to in the appended bibliography and detailed test data need not here be reproduced. However, one point of particular interest in connection with gages may be considered at this time, and that is whether file-hardness is concomitant with best wear resistance in any one steel.

The abrasive wear tests of Honda and Yamada (4) summarized in part in Fig. 3 indicate that martensitic (file-hard) high carbon steels have less resistance to wear than the same steels with troostitic structures (file-soft). As stated by these investigators, the wear of carbon steels "is greatest under the normalized condition, and becomes less in the sorbitic, troostitic, and martensitic

structure; but in high carbon steels the wear of martensite increases beyond that of troostite or sorbite." Similar results were obtained by Robin (21) who stated "the steel C 0.63, quenched in

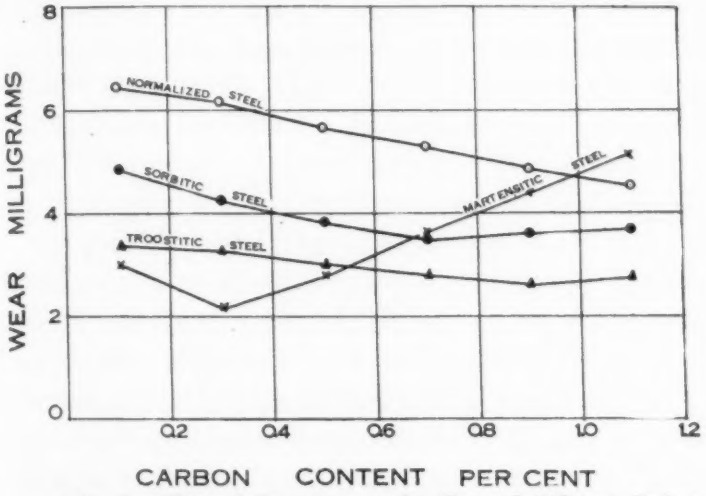


Fig. 3—Effect of Structure on the Wear of Different Carbon Steels as Determined by Honda and Yamada.
For Details of the Test Methods Employed, Refer to Footnote (4) of the Appended Bibliography.

oil, possesses, on ball and sclerometric tests, less hardness than the same steel when quenched in water, although nevertheless, its resistance to wear by emery is greater." While these and other results do not completely answer the question, they strongly indicate that file-hardness is not necessary for best wear resistance at least for some important conditions of service.

4. *Service factors affecting wear*

There are many factors arising in service, apart from the character of the metals involved, which have been shown in different investigations to have an effect upon wear. One of the most important is the contact pressure either between rolling or sliding metal surfaces or between a solid metal and finely divided material such as sand, emery, or other abrasives.

According to Robin (21) the "abrasion speed" may have a measurable effect upon the wear but under otherwise comparable conditions does not seem to be as important as the pressure. However, increase in either speed or pressure may change materially the temperature of the test specimens and the comparisons which are secured. As pointed out by Portevin and Nusbaumer (20) for bronzes under lubricated sliding friction, by Stanton and Batson

(15) for unlubricated rolling friction, etc., contact pressures may readily be sufficient to work-harden the surface of specimens and at least temporarily cut down the rate of wear.

In some cases the character of the surface is apparently of great importance in determining the rate and character of wear. One illustration will suffice: Jannin (13) states that "the principal cause of wear of bearing metal is insufficient polish."

Other factors which will at once be recognized as important are the presence or absence of a lubricating film and the nature, size and quantity of any abrasive particles present.

These and the many other variables cited in the foregoing discussion have been given mainly to show that the study of wear of metals is not only complicated by the wide variations encountered in service, and the many factors known or suspected to affect wear resistance but by lack of standardized methods of test and the difficulty or impossibility, at this time, of translating results into conditions differing from those under which any tests are carried out. In other words, there is not now sufficient information available concerning the fundamentals of wear testing so that a problem of wear resistance may be put on a routine test basis

III. Experimental Work

1. *Service wear tests of plug gages*

About 125 gages were prepared from different steels, heat treated in various ways, and submitted to a number of shops for tests under actual operating conditions but after about two years, less than half of this number had been returned with any information. Furthermore, careful examination of the performance records which were obtained showed that unless a sufficiently large number of additional gages could be prepared for wear in service, so that a statistical study could be made, little of the desired information could be secured. The variations in the conditions under which the gages were used, either in different shops or in gaging different parts in one shop made it impracticable to obtain desired comparisons and were suspected to have masked the effects of important variables. This is shown in the partial summary of performance records given in Table II.

Even without the details of the conditions under which the gages were used, it is evident that the differences in service con-

ditions, either at different plants (for example, columns 4 and 6, Table II) or in different "jobs" in one plant (for example, columns 3 and 4, Table II) were sufficient to produce wider variations in gage

Table II
Summary of the Service Wear Tests of Plug Gages

Steel	Gage Heat Treatment[2] (Temp. in °F.)	Holes gaged per 0.0001 inch wear on entering end at[1] Dodge Bros. (.875" gage)	Dodge Bros. (.750" gage)	White Motor Co.	Overland Motor Co.	Cadillac Motor Co.
1.10% C.....	1455/1470 w; 300	700	3880	1980	380	1200
1% C, 1.4% Cr.....	1470/1545 w; 300	860	2880	850	350	....
	1545 oil; 210/390	...	2850	850	380	1370
	1545/1560 oil; 300	...	1980	780	...	....
1.10% C, 1.35% W..	1470 w; 300	1080	....	...	...	....
Cr-Si-Mo steel[3]	1560 oil; 300	...	750	...	...	....
Stellite[4]	None	...	....	...	420	....
Chromium plated steel.	None	...	1080	...	...	....

[1]In general, the numerical values are the averages of results obtained on 2 gages in each case.

[2]1455/1470 w; 300 means water quenched from 1455 to 1470° F. and tempered at 300° F.

[3]1.2% C; 1.1% Si; 1.1% Cr; 1.0% Mo; 0.34% V; 0.3% W.

[4]Exact chemical composition not known; is an alloy of Co with metals of the Cr group, including Cr, W and Mo.

life than appreciable differences in the composition and heat treatment of the steels from which the gages were made.

It was therefore, considered advisable to carry out comparative tests under control in the laboratory planning subsequently to check the principal conclusions by a relatively small number of tests in service.

2. *The wear testing machine for plug gages*

The need for studying the wear of gages in the laboratory immediately brought up the question of a suitable testing machine. As has already been pointed out, wear tests must be carried out under conditions identical with or closely approximating those encountered in service if worthwhile results are to be secured. Since plug gages are ordinarily subjected to sliding abrasion, the machines of Saniter, Stanton and Batson, and Amsler, which give rolling friction, did not seem well adapted for the work in view.

The Derihon and Jannin testers, shown in Fig. 1, while primarily adapted to lubricated sliding friction might work successfully for dry wear tests, but would give data for only one set of conditions to which gages are ordinarily subjected. These machines cannot readily be adapted to a study of abrasive wear which would thus require an additional test unit.

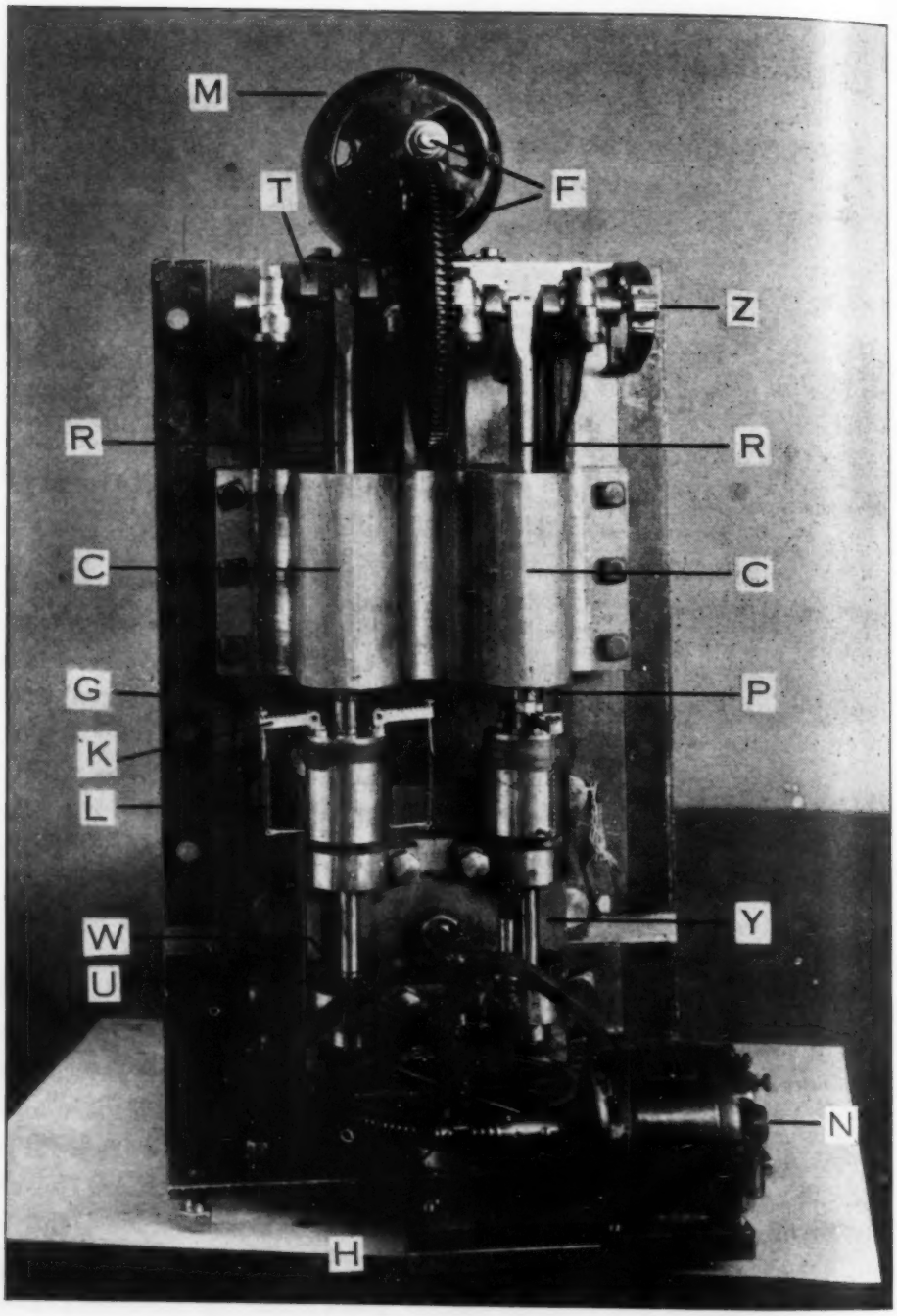


Fig. 4—The Author's Wear Testing Machine for Plug Gages.
Note that this Machine Provides for Simultaneous Tests of Two Gages.

While the machines for studying abrasive wear, as for example, those of Robin or Honda and Yamada, (Fig. 1), might be used successfully for wet or dry sliding friction (direct metal to

metal contact), difficulties would almost certainly be encountered in abrasive wear tests as it is not easy to secure uniform abrasive disks or papers.

In addition, none of the machines pictured in Fig. 1, appeared to approximate, to a desired degree, the conditions to which plug gages are most often subjected in service, and it was considered advisable to construct a special gage wear tester for the work in view.

As shown in Fig. 4, this is essentially a machine for repeatedly inserting plug gages into split rings, representing the material to be gaged, and permits control of a number of important variables such as the pressure between the gage and the work, the presence or absence of abrasives, lubricants, etc., and the metals from which both the gages and the material to be gaged are made.

The gage, G, (Fig. 4) is moved vertically in a split ring by means of a piston, P, sliding in a cylinder, C. The piston, which is driven by a motor, M, through reduction gears, F, a connecting rod, R, and crankshaft, T, is threaded at its lower end to provide means for attaching the test gage, G.

While the gage is being moved vertically, the split ring into which it has been inserted, is rotated horizontally by a motor, N, through reduction gears, H, a sprocket wheel, W, and chain, U. The movement of the gage combined with the rotation of the ring simulates the "wringing" action to which plug gages are frequently subjected in service.

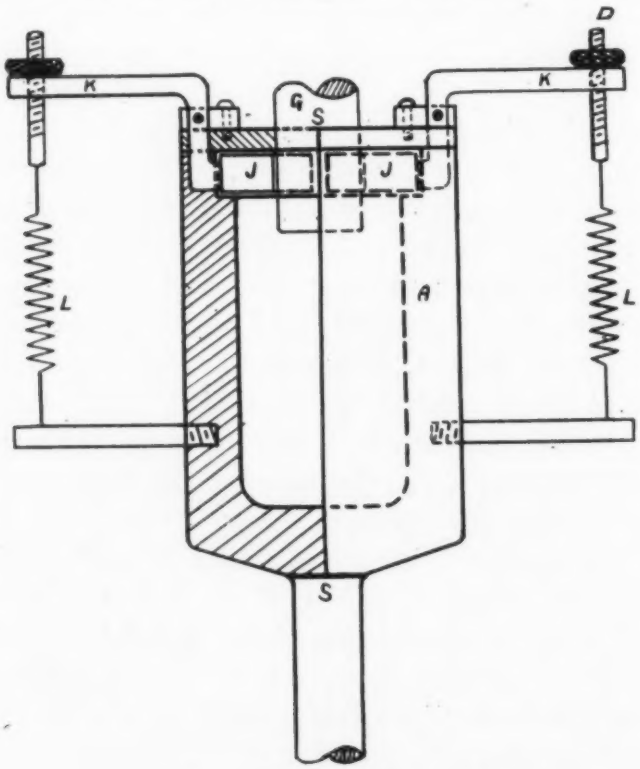
Control of contact pressure between the gage and the split ring is obtained by springs, L (Fig. 5), acting through lever arms, K. The holder for the split ring, J, is in the form of a cup, A, to provide means for studying wear in the presence of liquids.

The number of "gagings"[3] are recorded by a counter, Z, which is shown in Fig. 4. The machine is further provided with a number of adjustments including the length of stroke, alignment of gages and rings, etc. The cups holding the split rings are mounted on a carriage, Y (Fig. 1), which can be lowered when putting the test gages in the machine. With the gages in place, this carriage is raised to such a position that the bottom of the gage is about $\frac{1}{16}$ inch below the top of the ring when the piston is at the

[3] A "gaging" consists of a complete downward and upward movement of the gage in the ring, representing the insertion and removal of a gage in practice; it is equivalent to a 360-degree rotation of the crankshaft, T, Fig. 4.

top of its stroke. During a test, the gage is therefore never completely removed from the ring.

The springs, L, (Fig. 5), which control the contact pressures between the gage and the split ring, were calibrated in terms of threads of the adjusting screws, D. When the outside diameter



Sectioned Through SS.

Fig. 5—Sketch Showing Details of Ring Holder and Method of Controlling Contact Pressures in the Wear Tester.

of the split rings is kept constant there is no appreciable change in the position of the lever arms, K, and hence no cosine error which must be considered in transmitting the spring force to the gage and ring through the lever system.

3. *Methods employed with the gage wear tester*

The gages used in all tests had the form and dimensions shown in Fig. 6; they are cylinders 0.75 inches in diameter and 1.25 inches long with a hole at one end threaded to fit the pistons of the wear tester (P, Fig. 4). The split rings are likewise shown in Fig. 6 and have an outside diameter of 1.5 inch, a hole equal to 0.75 inches

and a height of 0.375 inch. All gages were ground and lapped but the rings were merely ground. With the gages and rings in place and the desired contact pressure between the two, the machine was

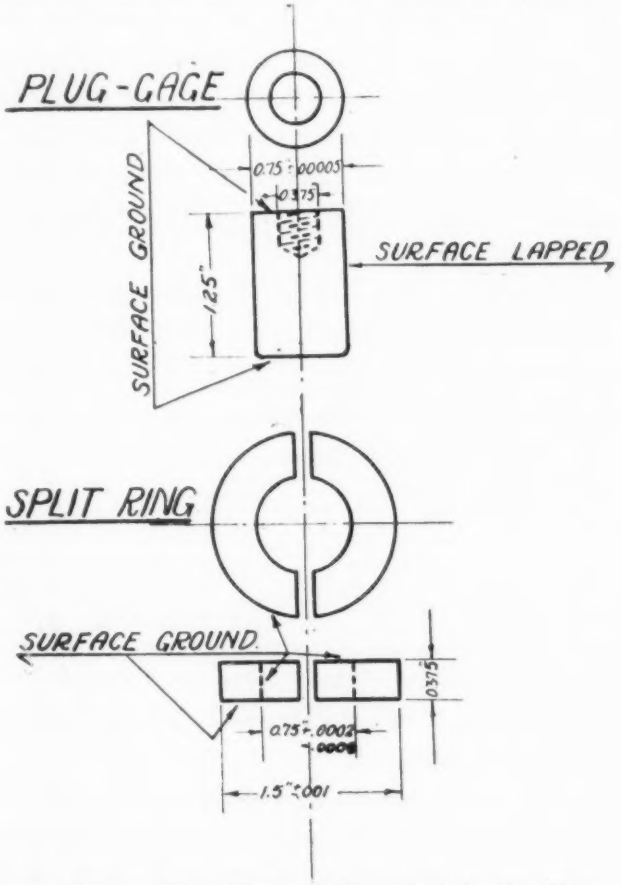


Fig. 6—Plug Gage and Ring Used in the Wear Tester Shown in Fig. 4.

started and run at a speed permitting 1200 to 1300 gagings, and a ring rotation of about 900 revolutions, per hour. After the desired number of gagings, the gages and rings were taken out of the machine and the wear determined by both decrease in diameter and loss in weight for the gage and loss in weight for the rings.

Loss in weight was determined to the nearest 0.1 milligram by weighing the gages and rings on an ordinary chemical balance, before and after each test. Change in diameter of the gage was determined from measurements on a Prestwich fluid gage made by the Coats Machine Tool Company.

4. *Steels used in the tests*

The different steels used and the heat treatments of the test

gages are summarized in Tables III and IV. In addition to carbon and a number of special tool steels, there are included two sets of chromium plated gages.[4]

The selection of the various steels and treatments summarized in Tables III and IV was dictated primarily by what had previously been done in the study of permanence and dimensional changes in hardening. By securing data on wear resistance for the same steels used in the studies of dimensional changes and permanence, a more complete picture is obtained of the value of the different metals for gage work.

Whenever possible the two heat treatments applied to each steel were selected to give respectively "file-hard" and "file-soft"

Table III
Chemical Compositions of the Steels Used in the Wear Tests

Steel Type	Mark	Chemical Composition, Per Cent C	Mn	P	S	Si	Cr	W	V	Remarks
1.05% carbon..	BC	1.06	.24	.014	.016	.18	....	.....	..	This steel used for all rings for the gage wear tester, for some gages and as a base for the chromium plated gages.
"Ball-race"	MB	1.01	.21	.016	.034	.29	1.46	.....	..	
"Oil hardening".	JJ	.86	1.17	.021	.010	.21	.51	.41	..	
"High speed"...	ZH	.75	.31	....	....	.11	3.92	17.95	.89	

conditions as one of the important questions raised by gage manufacturers was whether "file-hardness"[5] is necessary and a criterion of good wear resistance.

The chemical composition and heat treatment of the rings, representing the material to be gaged, were not varied. The rings used in all tests described in this report were "file-hard" and made from plain carbon tool steel containing about 1.05 per cent carbon (Steel BC, Table III) water-quenched from 1470 degrees Fahr. (800 degrees Cent.) and subsequently tempered at 300 degrees Fahr. (150 degrees Cent.). However, additional tests are now in progress in which the various gages will be used in contact with other metals such as commercial brass, aluminum casting alloys, soft steels and cast iron.

5. *Results obtained with the gage wear tester*

It was originally intended to make tests under two widely

[4]The details of the practice employed in plating the test gages with chromium are given in Appendix I. The plating was done under the supervision of W. Blum, Chemist, Bureau of Standards, by W. P. Barrows, Assistant Scientist.

[5]Resistance to penetration by a file or "file-hardness" is recognized as being dependent upon the type and quality of file employed as well as upon the manner in which the file is used; file-tests are, therefore, not exact. As used in this report, "file-hardness" means resistance to penetration by new fine Swiss type files, such as may be found in any machine shop.

Table IV

Results of Wear Tests Made With the Gage Wear Tester, Fig. IV, for Comparison of Different Steels and Heat Treatments

Gage Mark	Steel	Heat Treatment[a]	Diam. change per 4000 gagings in "metal" wear tests[b]—Inch×10⁴ 1	2	3	4	5	6	7	8	Av.	Holes gaged per 0.0001 Inch Wear	Diameter change per 2000 gagings in "abrasive" wear tests,[c] Inch×10⁴ 1	2	Av.	Holes gaged per 0.0001 Inch Wear	Gage Hardness Rockwell "C" Scale	File Yes or No
BC20	1.06 carbon	1470w; 300	.5	.2	.6	.2	.3	.3	.5	.3			2.1	3.0			65.9	yes
BC21	1.06 carbon	1470w; 300	.2	.2	.5	.4	.7	.1	.8	.3			2.0	2.6			66.3	yes
											.38	10,500			2.42	825		
BC12	1.06 carbon	1470 oil	0	0	.3	0	0	.2	..	.3			1.5	2.0			50.0	no
BC13	1.06 carbon	1470 oil	.1	0	.2	.1	0	0	.3	.4			1.9	2.5			46.7	no
											.13	30,800			2.00	1000		
MB1	"Ball race"	1470w; 300	0	.5	.2	.2	.3	.2	0	.3			3.5	1.7			62.9	yes
MB2	"Ball race"	1470w; 300	0	.4	.3	.3	0	.4	.4	.5			3.0	2.4			62.4	yes
											.25	16,000			2.65	755		
MB3	"Ball race"	1560 oil; 570	.2	..	0	0	.1	.3	.2	.2			1.8	3.4			58.6	no
MB4	"Ball race"	1560 oil; 570	.3	.1	0	0	.2	.3	.3	.2			2.5	3.4			58.6	no
											.16	25,000			2.78	720		
JJ1	"Oil-hardening"	1515 oil; 300	.3	.2	.3	.4	.1	.6	.2	0			2.4	2.4			64.1	yes
JJ2	"Oil-hardening"	1515 oil; 300	.2	.1	.4	.9	.1	.2	.5	.2			2.8	2.1			64.3	yes
											.30	13,300			2.40	830		
JJ7	"Oil-hardening"	1515 oil; 615	.2	0	0	0	.2	.1	.1	0			3.6	4.1			56.1	no
JJ8	"Oil-hardening"	1515 oil; 615	0	..	.3	0	.1	.3	.1	0			2.9	2.8			56.6	no
											.10	40,000			3.35	595		
ZH3	"High speed"	1650 furnace cool											1.7	1.8			22.3	no
ZH4	"High speed"	1650 furnace cool											1.9	1.8			22.8	no
															1.80	1110		
ZH1	"High speed"	2400 oil; 1100	.2	.2	.2	.1	.1	.2	.1	0			1.3	2.3			66.1	yes
ZH2	"High speed"	2400 oil; 1100	.2	0	.2	.3	.1	0	.4	.1			1.0	2.0			66.2	yes
											.15	26,000			1.65	1210		
BC14	Cr-plated	(on annealed steel)	0	0	0	0	.3	0	..	0			1.1	1.2				no
BC25	Cr-plated	(on annealed steel)	0	0	0	0	0	0	..	0			0.9	1.4				no
											.02	200,000			1.15	1740		
BC23	Cr-plated	(on medium hard steel)	0	0	0	0	0	0	.1	0			1.6	1.1				yes
BC24	Cr-plated	(on medium hard steel)	.1	0	0	0	0	0	.1	0			1.2	1.1				yes
											.02	200,000			1.25	1600		

[a] All temperatures given in °F. [b] Tests made with 4.5 lb. spring pressure. [c] Tests made with 1.5 lb. spring pressure.

different conditions representing extremes encountered by gages in service, namely, (1) wear under dry sliding friction in which one metal slides over the other, and (2) wear by abrasives. However, a number of difficulties were encountered which made it necessary to modify slightly these conditions of test.

With the gages sliding in direct contact (dry) with the split ring, even with very low contact pressures between the two, an

Fig. 7—Oxide Formation on One of the Gages Run "Dry" in Direct Contact with the Rings in the Gage Wear Tester.

oxide film was soon formed which protected both gage and ring from further wear or at least reduced appreciably the rate of wear. This oxide film, an example of which is shown in Fig. 7, may be ascribed to the fact that the small particles torn from the steel specimens are pyrophoric and oxidize the surfaces of the gages and rings. Apart from its cause, once a film was formed, a strong tendency was created for the gage and ring to "seize."

To avoid such difficulties, these tests were made in water, in which was dissolved approximately 1 per cent by weight of potassium dichromate ($\frac{1}{13}$ normal solution $K_2Cr_2O_7$) to minimize corrosion of the testing equipment, gages and rings. The presence of water is not believed to exert any appreciable influence upon the displacement and tearing off of the particles which produce and constitute wear, but serves primarily to prevent or minimize the tendency toward surface oxidation of the gages and rings. In any event, gages are frequently used in the presence of water, remaining from the coolant used in previous machining operations, and the

results should be equally as interesting and applicable as those from dry tests.

The first tests on wear by abrasives were made with oil-emery or water-emery mixtures placed in the cups forming the ring holders in the gage wear tester (marked A, in Fig. 5). While the abrasive mixture was agitated by the repeated immersion of the gage, the numerical values of wear were erratic. The replacement of the oil by a colloidal compound, such as Bentonite, was suggested to secure a better and more uniform suspension of the emery and hence more consistent and uniform wear. This, at first, appeared to offer a solution to the problem but not all lots of Bentonite used gave good suspensions and in addition there was a tendency to form a coating on the gage which was suspected of reducing the wear.

The procedure finally adopted consisted in feeding through a small copper tube directly upon the gages at a point just above the zone of contact with the ring, 25 grams of emery per liter of lard oil (approx. 3.4 ounces per gallon), kept under continuous and vigorous agitation and suspension in a supply tank by means of a propeller driven by a small electric motor. The supply tank was located a short distance from the feeding nozzle and the supply regulated by a valve. As will be evident from test data later discussed, consistent results were obtained with this method.

The two conditions of test to which reference has been made will hereafter be referred to respectively as "metal to metal" and "abrasive wear."

a. *File-hardness and wear*

The results given in Table IV throw further light upon the relation of file-hardness to wear. Both the 1.06 per cent carbon (BC) and chromium ball-race (MB) steels are more resistant to wear by one metal sliding over the other when in the file-soft conditions studied than when file-hard. In other words, with the size and type of gage employed, oil quenching, with or without the subsequent low temperature tempering, gives better wear resistance than water quenching. Expressed in terms of the number of holes gaged per ten thousandth of an inch wear on the entering end of the plug gage the file-soft high-carbon and chromium ball-race steels are about 2 to 3 times as resistant to wear, in these tests, as the same steels in the file-hard condition. However, no such

differences are observed in the "abrasive wear" tests in which the number of holes gaged per ten thousandth of an inch wear is more nearly the same for both the file-soft and file-hard conditions.

While the results of these "abrasive wear" tests do not confirm the conclusions reached by Honda and Yamada (4) or Robin (21) with respect to the wear of water quenched carbon steels, they are not to be considered unusual. Neither the file-hard nor the file-soft condition is proof against scratching or penetration by the emery used in the "abrasive wear" tests; the penetrating or "cutting power" of the emery is so high that the difference in toughness of the file-hard and file-soft conditions in these steels, while considerable, is undoubtedly not sufficient to make any great difference in the ease with which displaced particles are removed from the metal.

A different condition exists in the "metal to metal" wear tests where small particles must be torn off through the mutual sliding action between the gage and the ring. These detached particles, which themselves constitute an abrasive and cause more wear do not have the high "cutting power" of the emery and do not differ greatly whether torn from the file-hard or file-soft steels employed. Under such conditions, they are able to displace metal, but their effectiveness in detaching more particles is undoubtedly more largely a question of the toughness of the metal. As the file-soft steel is tougher than that which is file-hard the former wears less.

If this explanation is correct, similar effects should be observed in comparison of steel gages with chromium-plated gages for the chromium plating offers high resistance to penetration, but is brittle, as is evidenced by the frequency with which the coating "flakes" from the base metal. As shown in Table IV and Fig. 8, the chromium-plated steel gages show a remarkably long life in the "metal to metal" wear tests, but only a slight superiority in the "abrasive" wear tests in comparsion with the different steels tested. However, for ordinary gage work, in which both types of wear may be encountered, chromium plating appears to offer much promise.

b. *Comparisons of different steels and chromium plated metals for plug gages*

The principal difficulties with respect to the use of chromium-

coated gages are now associated with the mechanism of plating but, with the attention being given to chromium plating practice,

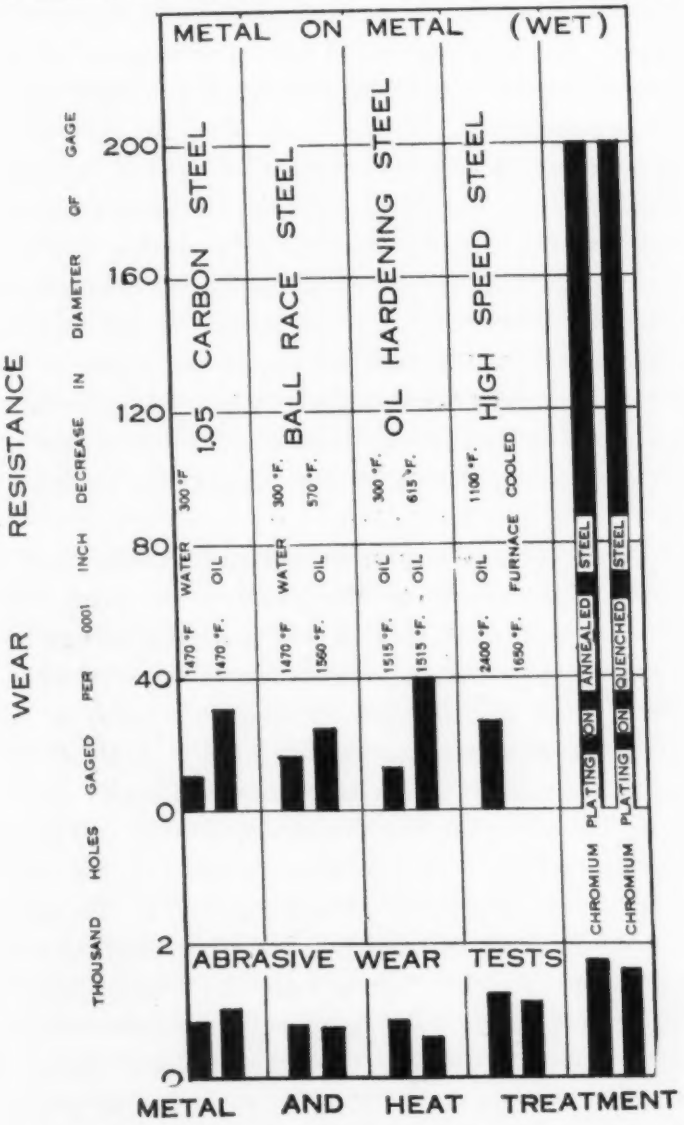


Fig. 8—Effect of Metal and Heat Treatment on the Wear Resistance of Plug Gages as Determined in Tests with the Gage Wear Tester, Fig. 4.

it should soon be possible to secure satisfactory coatings in a wide range of thicknesses on both simple and moderately complicated shapes without the "flaking" now frequently encountered.

The advantages of chromium-plated gages are not restricted to superiority in wear resistance under certain conditions of service. Apart from questions of cost an ideal gage metal should

(1) be readily machinable
(2) be free from dimensional changes on hardening
(3) have dimensional permanence in the hardened condition
and (4) have a high degree of wear resistance (includes sufficient indentation hardness so that it will not readily be scratched.)

Where chromium plating gives the desired wearing properties, as in some of the described tests, all of these requirements can be met because a soft metal base may be chosen which is readily machinable, which will have a high degree of dimensional permanence and which when coated with chromium to produce a hard good wearing surface will not be subject to the dimensional changes encountered in the heat treatment of steels. At present no other method is known by which all these requirements can be satisfied to the same degree as shown by the chromium plated gages here considered.

These comments are not to be taken to mean that chromium plating is the final answer to the problem of wear resistance of gages for there are cases known in which such platings have shown relatively poor performance. However, the described tests together with reports of practical experience[6] make it inadvisable to omit consideration of chromium plating for gages wherever poor performance is shown by the metals now ordinarily used.

Omitting for the moment consideration of chromium-plated gages, the most resistant to "abrasive wear" of the steels tested is the high-speed tool-steel but its superiority to the oil quenched and untempered 1.06 per cent carbon steel is practically negligible.

In the "metal to metal" wear tests the "oil-hardening" steel (JJ) tempered at a rather high temperature subsequent to hardening is the most resistant to wear. Of the file-hard steels, the hardened and tempered high speed steel is superior to the chromium ball-race steel which in turn is a little more resistant than the 1.06 per cent carbon steel. Further and more detailed comparisons are shown in Fig. 8 and Table IV.

It would appear from these tests that where abrasives are encountered, there is not much choice between the steels studied with the exception of the hardened and tempered high speed steel which is not quite as good as the chromium-plated gage. For un-

[6]Private communications.

lubricated sliding friction of one metal on another, chromium-plated gages are so far ahead of the different steels and treatments included in this investigation that it does not seem worthwhile to consider any of the latter.

It has already been pointed out that the original service wear tests, summarized in part in Table II, were not convincing, by themselves, on account of suspected variables in the service conditions and that for this reason, it seemed desirable to study wear under control in the laboratory. While the conditions of service cannot be exactly described sufficient information was submitted by the manufacturers co-operating in the service wear tests to indicate quite clearly that abrasives were present; it should therefore be of interest to compare the results of the "abrasive wear" tests in the gage tester with the results of the service wear tests (Table II).

When the results of the service wear tests are considered as a whole and minor differences in gage life disregarded, as they should be, it is seen from Table II, that the water quenched (and tempered) high carbon steel is just about as resistant to wear as the chromium ball race steel, either quenched in water (and tempered) or quenched in oil (and similiarly tempered). The same conclusions are reached by examining the results of the abrasive wear tests made with the gage tester and summarized in Table IV and Fig. 8.

Aside from numerical differences in gage life in the service wear tests the order of superiority of the several steels and treatments which give nearly comparable gage performance will vary depending upon which set of results are chosen as a basis of comparison but when considered as a whole show the same general results as were obtained from the wear tester. The single exception is found in the tests of the chromium-plated gages which, in the service wear tests, gave rather poor life. However, this can justly be ascribed to the fact that these gages were plated several years ago, at a time when chromium plating practice was not developed even to its present state. Even so, the results are significant for as has previously been indicated the exact conditions for which chromium plating shows best wear resistance are not now known. Its superiority for some conditions of service, however, warrants further consideration.

Insofar as comparisons can be made between the available service

wear tests and the tests with the gage wear testing machine, gratifying consistency has been observed.

c. *Relative wear of gages and rings*

One of the most interesting features revealed by the described wear tests is shown in comparisons of the loss in weight of the gages and rings. Examination of the results of the "abrasive wear" tests included in Table V. shows that when a relatively soft gage is

Table V
Weight Losses of Gages and Rings in the Tests Summarized in Table IV

Gage		Average weight loss per 4000 gagings in "metal to metal" wear tests, mg.[b]		Average weight loss per 2000 gagings in "abrasive" wear tests, mg.[c]	
Steel	Heat Treatment[a]	Gage	Ring	Gage	Ring
1.06 carbon ...	1470 w; 300	24.9	19.3	4.7	2.3
1.06 carbon ...	1470 oil	23.4	80.2	1.7	4.6
"Ball race" ...	1470 w; 300	33.4	33.8	6.2	3.8
"Ball race" ...	1560 oil; 570	31.6	46.0	2.0	3.8
"Oil-hardening".	1515 oil; 300	33.4	24.3	4.8	2.2
"Oil-hardening".	1515 oil; 615	37.1	57.1	1.9	4.3
"High speed"...	1650 furnace cool	21.3	86.2	...	...
"High speed"...	2400 oil; 1100	21.2	17.0	3.5	4.7
Cr-plated	on soft steel	12.9	29.9	1.4	1.8
Cr-plated	on med. hard steel	18.6	25.5	1.0	2.0

[a]All temperatures given in °F.; thus 1470 w; 300 means water quenched from 1470° F. and tempered at 300° F.

[b]Tests made with a spring pressure of 4.5 lbs., results are the average of 8 tests on each of 2 gages.

[c]Tests made with a spring pressure of 1.5 lbs., results are the average of 2 tests on each of 2 gages.

used opposite a hard ring, the wear of the latter is much greater than that in the former. This condition is found with the oil quenched carbon steel, the annealed high speed steel, the "oil hardening" steel when tempered at a high temperature subsequent to hardening, and, to a smaller degree, in the oil-quenched and tempered ball-race steel.

This effect cannot be considered unusual as it is found with a variety of steels. A probable explanation is that the relatively soft gages become "charged" with the abrasive, which at least to some degree, protects them from rapid wear. At the same time, this "charging" promotes concentration of the abrasive and the soft gage becomes in reality an "emery lap" which more quickly scratches, penetrates or displaces small particles from the ring. Since the ring is file hard, and comparatively brittle, the displaced particles are readily torn from the surface of the ring and the net result is greatly accelerated wear.

If this explanation is correct, a similar effect should be ob-

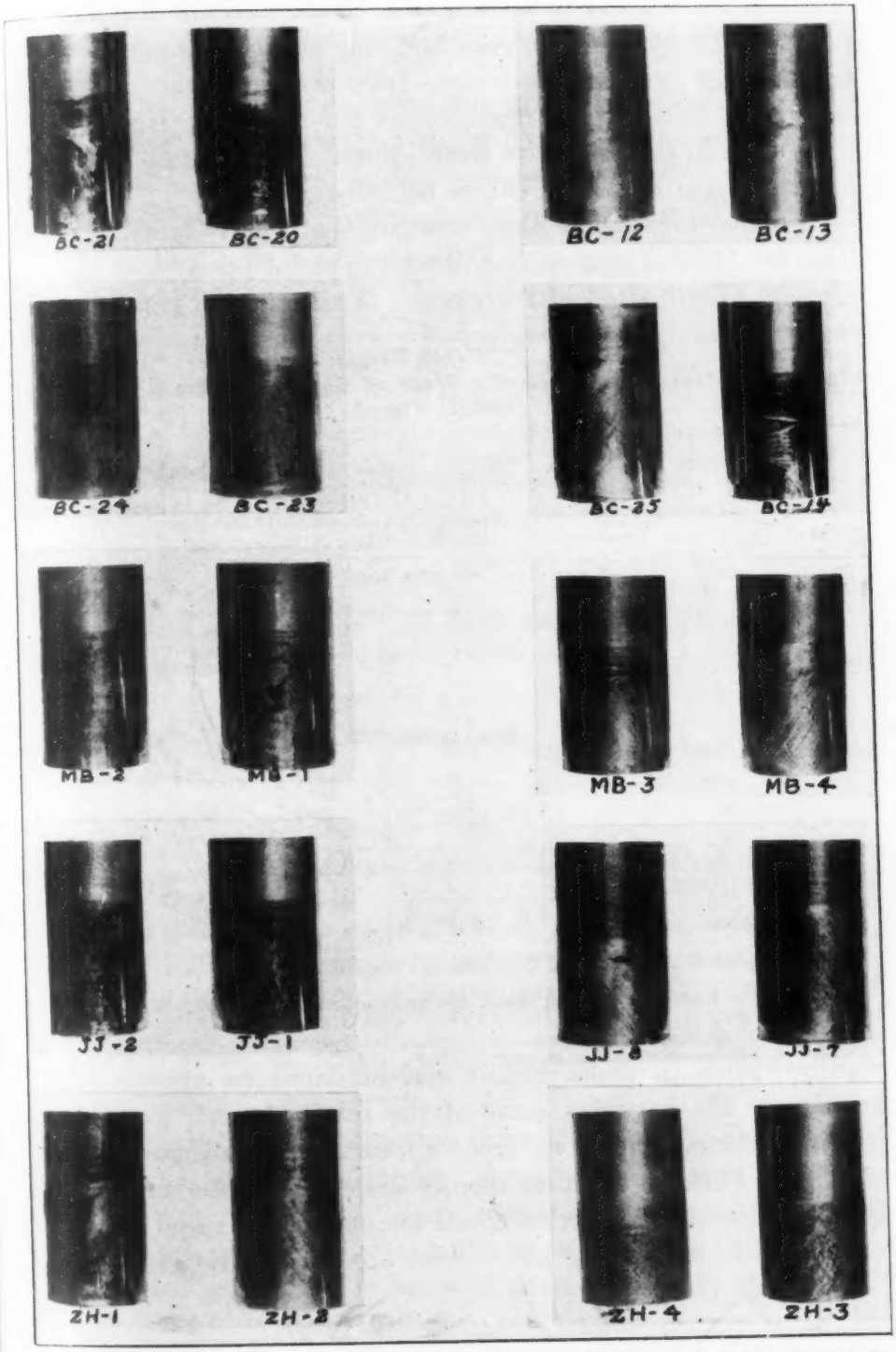


Fig. 9—Photograph Showing the Appearance of Different Steels Heat Treated in Various Ways, After "Metal to Metal" Wear Tests in the Gage Wear Tester, Fig. 4.

served in the "metal to metal" wear tests, but the differences in loss of weight between the gage and ring should be smaller because the abrading particles must come from the gages and rings themselves, and thus have a much lower penetrating power than the emery used in the "abrasive wear" tests. Examination of Table V shows that the suspected effects actually exist.

Further evidence of the "charging" of relatively soft gages in either the "abrasive wear" or "metal to metal" wear tests is found in visual examination of the gages. While the "charging" is not

Table VI
Effect of Contact Pressure on the Wear of Gages and Rings in the Wear Tester, Fig. 4

Gage		Gage Wear Per 1000 Holes gaged as measured by[b]			
Steel	Heat Treatment	Contact Pressure,[a] Lb./In.^{2}	Diameter Decrease, Inch × 10^4	Wgt. Loss, mg.	Corresponding Ring Wear,[b] mg.
		"Abrasive Wear" Tests			
1.06 carbon ..	1470° F. w; 300° F.	0[d]	.35	6.4	5.9
		8	1.1	12.5	9.6
		16	1.3	19.6	16.7
		27	2.9	24.0	18.7
1.06 carbon ..	1470° F. oil	0[d]	.85	6.2	9.5
		8	1.0	11.7	40.1
		16	1.8	18.6	47.8
		27	1.8	17.5	57.4
		"Metal to Metal" Wear Tests			
1.06 carbon ..	1470° F. w; 300° F.	0[d]	.07[c]	....	....
		8	.13[c]	....	....
		16	.18[c]	....	....
		27	.30[c]	....	....
		35	.63[c]	....	....

[a] Calculated from spring pressures on the gage wear tester, from the equation

$$\frac{\text{Spring pressure} \times \text{lever arm ratio}}{\text{contact area}} = \text{Spring pressure} \times 5.4.$$

[b] Except as otherwise indicated, the values given are the averages of 2 tests on each of 2 gages.

[c] These represent wear per 2000 holes gaged; they are the average of 3 tests on each of 2 gages.

[d] This is the least pressure with which the machine can be operated; it is not actually zero but very low.

readily visible in photographic reproductions on account of the fineness of the particles, some of the relatively soft gages shown in Fig. 9 clearly have a different appearance than those which are file hard. Perhaps the most readily discernible difference is shown in comparison of the hardened and tempered high speed steel gages (ZH1 and 2) with those which were annealed (ZH4 and 5).

d. *Pressure versus wear*

To throw further light upon the wear resistance of the different steels and treatments, several sets of experiments were made

in which the contact pressure between the gages and rings was varied. The results obtained are given in Table VI and summarized in Fig. 10.

Within the limits investigated, the wear of the file-hard carbon steel gages and rings increases with the contact pressure in both the

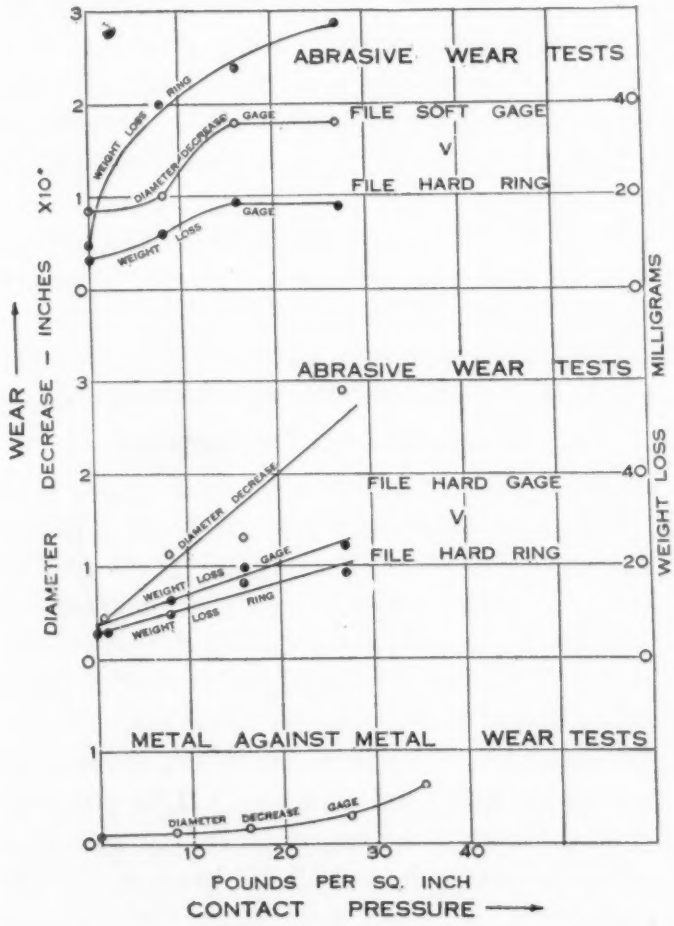


Fig. 10—Effect of Contact Pressure on the Wear of Gages and Rings in the Gage Wear Tester.

"metal to metal" and "abrasive wear" tests. The wear of the file-soft carbon steel gages in the "abrasive wear" tests at first increases, but then reaches a value which is not raised by further increase in the pressure between the gage and ring.

The fact that, under increase in contact pressure above a certain value, the wear of the file-soft steel does not increase while that of the file-hard steel does, is quite consistent with comparisons made earlier in this report, and may be explained by the "charging" of

the file-soft gage. This becomes, in effect, a "lap" which increases the rate of wear of the file-hard ring.

However, of greater significance, is the fact that the amount and rate of wear produced by an abrasive such as emery is not only affected by contact pressures, (clearance between the gage and the part being gaged) and the properties of the gage, but also by the properties of the material being gaged (the "work"). Therefore, no attempt should be made to extend the results of the described tests to the gaging of metals other than file-hard steels similar to those here used without first obtaining suitable experimental data.

However, with the form of the pressure-wear curves shown in Fig. 10, it may be said that change in pressure from that used in the comparisons of the different steels and treatments for gages would alter the numerical values of wear but not the order of superiority where appreciable differences originally existed.

IV. Summary and Conclusions

1. A new wear testing machine for plug gages is described.

2. With this gage wear tester a study was made of the wear of heat treated steel gages and chromium-plated steel gages (1) under wet sliding friction between two solid metals ("metal to metal" wear), and (2) in contact with finely divided emery in lard oil ("abrasive wear" tests). The comparisons obtained from the one type of test differed from those in the other.

3. Under wet sliding friction in which the various gages were in contact with water quenched (file-hard) 1.06 per cent carbon steel rings the chromium-plated gages showed an exceedingly high resistance to wear in comparison with file-hard or file-soft carbon and special tool steels. File-soft steel gages were more resistant to wear than gages made from the same steels in the file-hard condition as measured either by loss in weight or reduction in diameter.

4. The chromium-plated gages did not show the marked superiority in wear resistance in the "abrasive wear" tests which was observed in "metal to metal" wear test. Their life was more nearly the same as gages made from quenched and tempered high speed tool steel which, in turn, was somewhat more resistant to wear than file-hard or file-soft high carbon, chromium ball-race and so-called "oil hardening—non deforming" tool steels.

5. In the "abrasive wear" tests, and to a lesser degree in the

"metal to metal" wear tests, the "soft" gages wore less than the file-hard high carbon steel rings in which they were sliding. This was ascribed to the "charging" of the relatively soft gages with emery or particles of metal torn from the gages and rings which thus protected the soft gages from further rapid wear. At the same time this "charging" promoted concentration of abrasive material and the soft gages became in reality a "lap" which caused rapid wear of the file-hard rings.

6. The results of the "abrasive wear" tests made with the gage wear testing machine were found to be comparable to results of a few service wear tests on similar materials, in which the gages encountered abrasives of a similar type.

Acknowledgements

Acknowledgement is made to D. R. Miller and assistants of the Gage Section, Bureau of Standards, for assistance in measuring many of the test gages and for suggestions and general co-operation; to F. C. Raab, laboratory mechanic, who built the gage wear testing machine devised by the authors; to H. W. Gillett, chief, division of metallurgy, for suggesting the inclusion of chromium-plated gages in this investigation and other helpful suggestions; to H. W. Bearce, co-chief of the division of weights and measures, Bureau of Standards; and Howard Scott, formerly investigator for the Gage Steel Committee. who arranged for the service tests of the gages.

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VI. Appendix I. Methods Used in Chromium Plating the Test Gages

(See footnote (4) of the text)

The conditions used for plating the gages were as follows: (A more detailed description of the chromium plating process employed is given by H. E. Haring, *Chemical and Metallurgical Engineering*, 32, 692, 1925; and also in Letter Circular 177 of the Bureau of Standards.)

Solution Composition	grams per Liter
Chromic acid, CrO_3	250
Chromium sulphate, $Cr_2(SO_4)_3$	3
Chromium carbonate, $Cr_2O(CO_3)_2$	7

The gages were plated one at a time in a small rectangular glass jar, the dimensions of which were approximately 8 x 8 centimeters square and 13 centimeters deep, and which contained about 600 cubic centimeters of the above solution. Anodes of sheet lead about 1 centimeter wide and 0.1 centimeter thick were suspended in each of the four corners of the jar to a depth of approximately 2 centimeters. The solution was maintained at a temperature of 45 degrees ± 2 degrees Cent. during the plating operation. The gage to be plated was well scrubbed with pumice and water, dried with alcohol and ether, and was then suspended in a vertical position in the center of the jar. Connections were so made that the introduction of the gage into the solution closed the plating circuit. An average current density of 10 amperes per square decimeter (93 amperes per square foot) was maintained on the cathode and deposition was continued for six hours. The deposit obtained was radially uniform, although it was somewhat thicker on the bottom edge. On no part of the gage was the deposit less than 0.02 millimeters (0.0008 inches) thick at the completion of the plating operation.

The Authors

H. J. French, born in New York City, attended a private school and subsequently the School of Mines of Columbia University, from which he was graduated with the degree of Metallurgical Engineer. Following practical experience in the West, he became associated

H. J. FRENCH

H. K. HERSCHMAN

with an Eastern automobile and airplane engine plant as Metallurgical Engineer. During the World War he was engaged in metallurgical work for the Bureau of Aircraft Production, and in 1919, joined the metallurgical staff of the Bureau of Standards.

He is the author of numerous technical articles, a member of Sigma Xi, and several years ago was chairman of the Washington Chapter of the American Society for Steel Treating.

H. K. Herschman was born in Washington, D. C., in 1899, and educated in the public schools of that city. He entered the photometric section of the Bureau of Standards in 1916, leaving the Bureau in 1917 to enter the University of Michigan, from which institution he received the degree of Bachelor of Science in Chemical Engineering in 1921. Mr. Herschman returned to the Bureau of Standards in 1921, where he has since been engaged in research metallurgy. He received the degree of Bachelor of Law in 1925 from George Washington University Law School and became a member of the Bar of the District of Columbia in 1926.

DISCUSSION

Written Discussion: By B. H. Blood, Hartford, Connecticut.

The present paper performs a real service to industry by emphasizing the complexity of the problems which the authors have undertaken to solve, and the necessity for cooperation from gage users if laboratory results are to be checked against actual performance. It is fully apparent to all who have worked on these problems that no one steel and no one heat treatment is likely to be found best for all conditions. Assuming that gages could be exactly duplicated in all respects, the wearing action is dependent upon such variables as different metals and different structures of metals gaged, conditions of lubrication, presence of abrasive, smoothness of holes gaged, pressures used in gaging, and the personal equation of the operator. Persistent analytical work such as the authors are doing, carefully checked by service tests in different shops, may be expected to increase our present knowledge and improve our shop practice.

If practical tests in user's shops are to be of their greatest value they should be carefully recorded as to all possible variables, and the worn out gages returned as promptly as may be, with full information, to the Bureau of Standards. In this way comparisons can be made and discrepancies checked or reconciled. It will be a longer and more complex investigation than seemed possible when the Gage Steel Committee was organized. The authors of this paper have shown the way to valuable results ahead, and it is to be hoped that money and time, as well as general interest and cooperation, will be available to carry it through.

The testing machine described seems to be capable of reproducing service conditions much more closely and under better control than anything previously available. In fact, wear tests which promised to be comparable with service tests have until now been sought without success. As the working surface of the split ring is quickly changed from the original ground surface by burnishing and lapping, it does not exactly reproduce service conditions, but the results will probably be comparable.

For the "abrasive wear" tests it may be that, instead of lard oil and emery, shop conditions might be more closely simulated by the use of a water solution such as is generally employed in grinding operations, carrying the usual amount of abrasive and steel chips. Lard oil and emery mixtures are not often met with in gaging holes.

While plug gages are the logical starting point for this investigation, similar work on thread gages may be of greater importance commercially. Many conditions will be different, and the same gage metal will probably not prove best for both. Here difficulties in gage fabrication will have to be considered. The work will usually be soft, and abrasive will not be an important factor. The field for research is broad, but the results will be correspondingly valuable.

In view of the results already shown, the wisdom of requiring gages to be file hard may be further questioned, though this may depend upon the metal to be gaged. The original finish of the gage is found to be quite important, as a highly finished gage resists scratching, but when

fine scratches begin to form they gather abrasive particles and extend themselves quite rapidly.

The common practice of gaging ground holes without first removing coolant and grit seems rather brutal from the gage maker's point of view, but the higher gage cost resulting is probably justified by the time saved in production.

It will, I believe, be found worth while to use one steel or temper for working gages and another for final inspection where the work has been cleaned. We may find a choice between gage steels best adapted to resist wear in ground holes, depending upon whether the work is file hard or somewhat drawn, and also between reamed holes and ground holes.

Freak steels, and freak specimens of the same steels, complicate the problem and emphasize the necessity for check tests under service conditions. Occasionally a gage has been found to give several times the average life of the lot from which it was taken. Such examples when found are worthy of careful study. Any results from this investigation which teach the gage maker to produce and the gage user to demand more serviceable gages will be mutually helpful. The burden, as well as the benefit, should be shared between them.

Written Discussion: By L. A. Danse, Cadillac Motor Car Company, Detroit.

After going over the paper with O. J. Snider, supervisor of tool inspection, who is a member of the Gage Steel Committee, the following comments are in order, most of them originating with Mr. Snider:

The machine developed for testing gages is an ingenious design and marks a distinct step forward in the project. However, additional work is necessary in view of the fact that Cadillac has found that tests run by repeated applications of the same test surfaces to the gage do not wear the gage nearly so rapidly as do applications of new surfaces to the gage. We are not sure what the ratio would be, but it would doubtless be shown that new surfaces would wear the gage twice as fast as re-applications of the same surface.

Photographic study of the surfaces applied to the gage shows that a re-applied surface is lapped off at each application, from the start; and after a few hundred applications to the gage, the re-applied surface has entirely lost its "teeth," that is, the roughness left by the grinding wheel.

On the other hand, where new surfaces are applied to the gage at each gaging stroke, the grinding wheel roughness is always fresh and sharp. This surely has a considerable effect in increasing wear.

It is suggested that changing of the split ring wearing surfaces at short intervals, to more nearly simulate actual service conditions would be an improvement. The rings could be re-ground, to the same internal radius, several times before they became useless.

Another condition which we have found to exercise a definite effect is the complete removal of the gage at each application of the wearing surface. If the gage is not completely removed, the wear is less than when

the gage is removed entirely and inserted in a slightly different position.

In view of Cadillac findings as above outlined, it is to be noted that our tests positively demonstrate the superiority of file hard gages over file soft ones.

A gage with a soft surface will pick up small particles from the wearing surface. These particles will "load" onto the surface of the gage and continue to pick up until they grow large enough to break loose and scratch their way out of contact.

Due to this and other causes, gages which are soft to a file are not uniform in wearing qualities, even though the hardness over the contact surfaces be uniform.

Gages which are file hard apparently wear about the same, in our service tests.

Other peculiarities worthy of remark are that high speed steel plug gages wear along the whole contact surface uniformly. Carbon steel or Stellite wears worse on the entering end of the gage, over about 1/8 or 3/16 inch of the length of the contacting surface.

It has been noted that with Stellite gages, "loading" is seldom encountered, even though the Stellite may not be file hard. The gage has an entirely different "feel" which is easily noticed.

Chromium-plated gages also have a marked resistance to "loading" and a different "feel".

Oral Discussion

A. H. D'ARCAMBAL: I have read Messrs. French's and Herschman's paper with considerable interest, being a member of the Gage Steel Committee.

The wear test machine constructed by the authors should give reliable results, as the gages tested in this novel machine seemed to be working under the same conditions as gages in actual service. The results obtained, however, do not check with the results obtained from gages in service.

In the first place, plug gages made of plain carbon tool steel and incorrectly hardened by oil quenching would give very little service under actual working conditions for the following reasons:

First, these gages would not take such a high polish as file hard gages and, therefore, have a shorter life.

Second, one cannot obtain the proper feel with gages possessing a Rockwell hardness of only 46.7 C to 50 C.

Third, gages as soft as these would be easily bruised, thereby ruining the same.

The results obtained by the authors show that these file soft gages gaged three times as many holes—metal to metal test—and 25 per cent more holes in the so-called abrasive test. What the gage manufacturers are primarily interested in is the comparative wearing quality of file hard gages with gages correctly hardened but drawn say to 425 or 450 degrees Fahr. instead of the usual 350 degrees Fahr. draw.

It is difficult to believe that annealed high speed steel gages would

give approximately the same service as high speed steel gages properly hardened and tempered. One could not obtain a satisfactory finish on gages made from annealed high speed steel, and the "feel" factor, bruising, etc., would also work against these very soft steel gages.

The results obtained by the authors on chromium-plated gages were very interesting. Service results check with the ones obtained by the wear test machine in the metal to metal wear test. In the abrasive wear test, however, the chromium-plated gages perform only slightly better than carbon tool steel gages. Much depends on the procedure used for chromium plating, etc. Moreover, some investigators suggest using a hardened tool steel gage for plating, so as to have more of a backing.

Years of experience in manufacturing plug gages has taught us that the majority of gage users require a gage that is either file hard or one that can just be touched with a temper testing file.

We would suggest that the authors continue their experiments with the wear testing machine on carbon and alloy tool steel gages that have been correctly hardened and then drawn about 50 degrees Fahr. higher than the highest temperature that will produce a file hard gage. The results should be interesting.

O. Z. Klopsch: We have experimented with chromium plated gages, that is, chromium plating on case hardened cold-rolled steel, and our results showed a remarkably close check to the results the authors obtained. A case hardened gage may in this case be considered as about the same as the hardened 1.05 per cent carbon steel hardened that the authors used. In our experiments we obtained four to eight times increased wear on the plug gages using them in cylinder bores. The cylinder bores were lapped with kerosene and emery so there probably was a little of the abrasive present which from the results obtained evidently had little effect.

E. J. Bryant: There is no question but that the wear of gages is a very important one to all of us. The steels used are an important factor. The one thing that seems to be of as much importance as any of these in the ultimate result obtained from the gage is the finish on the gage itself when it is put to work. From the gages that we have put out and from the tests that I have had reports of, the finish on the gage itself pretty nearly determines the life of the gage on work where the limits are close, and, of course, that is where we get the large gage use; where the producers have holes to produce within five ten-thousandths or one-thousandth of an inch limit, the gages must be very nearly right in order to give the operator a reasonable working tolerance and consequently the gage must give full wear value in order to keep the cost down.

So that the finish on the gage surface is one of the points that has to be taken into consideration in the final results in checking gage wear. Of course, as Mr. d'Arcambal has pointed out, in order to obtain that finish one must have the proper degree of hardness. We cannot go down to a low degree of hardness and obtain the proper finish. Our experiments have checked very closely with Mr. d'Arcambal's. Best results are obtained with a steel hardened just below the file hard state, where the file just begins to pick up and then producing a good finish on that gage.

I am sure that the work of the Gage Steel Committee, that is, the work of the Bureau of Standards, is going to be of big benefit to all of us and we owe them much. Before the tests are completed we will find out new factors that may help us considerably.

H. J. FRENCH: Messrs. Blood, d'Arcambal, and Bryant have each commented on the importance of the character of finish on the life of the gage. We fully recognize that, and, as given in the report, have carefully recorded the conditions of preparation which were in reality "standardized" for all the gages used in these tests.

I think I can best illustrate the importance of the surface condition of the gage by citing the case of the chromium plated gages in the metal to metal wear tests. In the chart which was thrown on the screen a few moments ago, a marked superiority in wear resistance was found for the chromium plated gages. Now these gages were carefully lapped after plating; when the same plating process was employed and the unlapped gage was tested, poor performance was found, which, I think, is a very good example of the importance of the character of finish on the gage.

Mr. d'Arcambal's comments refer primarily to the application of steels rather than to their wearing properties. We are not recommending in this report the use of file soft gages, but are simply reporting the results of experiments which show that file soft steels may have a better wear resistance than the same steels in the file hard condition. We do not consider that we will be in a position to recommend applications of steels or treatments for resisting gage wear until we have done much further work.

I would like to call Mr. d'Arcambal's attention to the fact that we were unable, as shown in the report, to use the annealed high speed steel in the metal to metal wear test.

With reference to Mr. d'Arcambal's question on the wear of file soft gages tempered at high temperatures but which were quenched in the same manner as those which were file hard, I would call attention to data for the "oil-hardening" steel given in the report. The same coolant was used in hardening the different gages made from this steel and file softness obtained by tempering. Note that the wear of the file soft gages was less than that of the gages which were file hard. I believe this answers, at least in part, the question raised.

Mr. Danse has suggested that a wear test in which the rings were more frequently replaced would more nearly approximate practical service as it would serve to keep the gage throughout the test in contact with the type of finish ordinarily encountered in service. His point is a good one and originally received considerable attention from the authors. Aside from the probability of more erratic results in the laboratory such a test would be more nearly a test of the surface than a test of the metal and heat treatment used in the preparation of the gage and so was discarded.

It should be kept in mind that what is desired is to place in proper order of wear resistance different steels (and treatments) and for this it is not necessary to secure numerical values of wear which approximate those of prac-

(Continued on Page 813)

THE HARDENING OF METALS BY DISPERSED CONSTITUENTS PRECIPITATED FROM SOLID SOLUTIONS

By Robert S. Archer

Abstract

Metals may be effectively hardened by highly dispersed particles within the grains, such as may be formed from supersaturated solid solutions. The typical process of hardening by this means consists in a solution heat treatment at a relatively high temperature followed by rapid cooling into a region of supersaturation, then by a precipitation treatment or aging to permit the formation of a very fine precipitate. The precipitate tends to grow through a critical particle size conducive to maximum hardness, beyond which the hardness decreases.

Various examples of this type of hardening are described. Some generalizations regarding the theory of this process are offered.

THE idea that the hardness of an alloy may be increased by the precipitation of a soluble constituent from solid solution was first advanced by Merica[1] and his associates in a hypothesis to account for the age-hardening of duralumin. It had been generally held that the formation of a metallic solid solution from an aggregate caused increased hardness and, conversely, that the decomposition of a solid solution involved decreased hardness. Merica's theory of the duralumin phenomena was in fact based on this general proposition in that the hardening of annealed duralumin (an aggregate) by heating and quenching was attributed to the formation of a solid solution, but on an exception to the general proposition in that the further hardening of the quenched alloy on "aging" at ordinary temperatures was attributed to the decomposition of this solid solution to form an aggregate in which one of the constituents was very highly dispersed.

This new idea has proved very fruitful not only in the clarification of known metallurgical processes, but in the development of new ones. The announcement of this explanation of the hardening

[1]Merica, P. D., Waltenberg, R. G., and Scott, H., Heat Treatment of Duralumin, *Bulletin*, American Institute of Mining and Metallurgical Engineers, June, 1919. Also U. S. Bureau of Standards Scientific Paper No. 347, November 15, 1919.

The author, R. S. Archer, Research Bureau, Aluminum Company of America, Cleveland.

of duralumin has been followed by the discovery of analogous processes for hardening such diverse metals as magnesium, copper, lead and iron. It is proposed in this paper to describe some of these subsequent discoveries and in connection therewith to develop the general theory of what may be called "precipitation hardening."

Aluminum-Copper Alloys

Aluminum and copper form a compound $CuAl_2$ containing about 54 per cent copper. A eutectic occurs at 33 per cent copper,

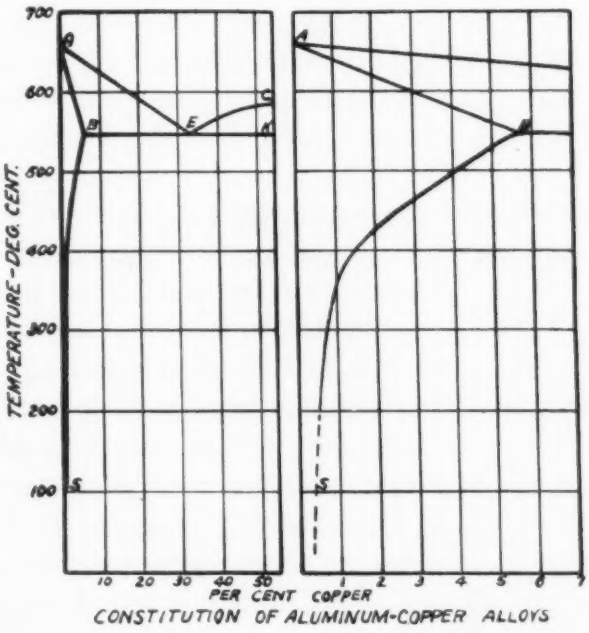


Fig. 1—Constitution of Aluminum-Copper Alloys (Dix and Richardson)

the eutectic mixture consisting of the compound $CuAl_2$ and a solid solution of copper in aluminum. The solubility of copper in solid aluminum decreases from about 5.5 per cent at the eutectic temperature, 548 degrees Cent., to less than 0.5 per cent at room temperature, as shown by the solubility curve SB in the constitution diagram Fig. 1.[2]

Consider, for example, an alloy containing 4.5 per cent copper. By prolonged heating at 968 degrees Fahr. (520 degrees Cent.) or above, all of the copper can be brought into solid solu-

[2]Dix, E. H., Jr., and Richardson, H. H., Equilibrium Relations in Aluminum-Copper Alloys of High Purity, Paper of American Institute of Mining and Metallurgical Engineers, No. 1534-E, issued with *Mining and Metallurgy*, February, 1926.

tion. This solid solution can be retained at ordinary temperatures in a supersaturated condition by quenching in water. It appears that some precipitation of $CuAl_2$ from this supersaturated solution takes place spontaneously on standing at room temperature but the precipitation is much more marked if the alloy is reheated at sightly elevated temperatures. This reheating, called "artificial aging", is generally carried out at about 257 to 320 degrees Fahr. (125 to 160 degrees Cent.), and is a precipitation process. The precipitates formed at these temperatures are too fine for resolution under the microscope, their presence being inferred from indirect evidence. On heating at higher temperatures or for longer periods of time, however, precipitation is followed by an increase in the average particle size and precipitates formed at 392 to 662 degrees Fahr. (200 to 350 degrees Cent.) have been observed under the microscope.

This alloy reaches its maximum hardness and minimum plasticity when the copper is precipitated in the form of very small particles of $CuAl_2$. These particles are too small for direct observation but have been estimated to have average diameters on the order of 0.0000001—0.000001 inches. As the particle size increases, the hardness of the alloy decreases continuously and the plasticity increases. With an average particle diameter of about 0.0001 inches the alloy is in the annealed condition and is extremely soft. The solid solution produced by quenching from the high temperature is intermediate in hardness between these extremes. In this solid solution it is probable that the copper is atomically dispersed.

As an illustration of the effect of the manner in which the copper is distributed, the following approximate values are given for the Brinell hardness of a pure alloy containing 4.5 per cent copper in various conditions.

Condition	Brinell Hardness
Annealed — large particles of $CuAl_2$	30
Quenched from 540° C. — solid solution	60
Quenched from 540° C. — artificially aged at 150° C. $CuAl_2$ precipitated as very fine particles	110

Duralumin

The hardening of aluminum alloys by heat treatment began with the investigations of Alfred Wilm (1903-1911). The word "duralumin" being a trade name has not an exact metallurgical

meaning, but is generally understood to apply to a class of wrought aluminum alloys containing both copper and magnesium. A typical composition is: copper 4.0 per cent; magnesium 0.5 per cent; manganese 0.5 per cent; remainder commercial aluminum containing say 0.5 per cent iron and 0.3 per cent silicon as impurities. After hot rolling, cold rolling and annealing at about 617 degrees Fahr. (325 degrees Cent.), this alloy has a tensile strength of about 27,000 pounds per square inch, an elongation of about 16 per cent, and a Brinell hardness of about 50. The hardening treatment consists in heating to about 932 degrees Fahr. (500 degrees Cent.) and quenching. Immediately after the quenching, the tensile strength may be about 45,000 pounds per square inch, the elongation about 20 per cent, and the hardness about 70. As the material stands or "ages" at room temperature a spontaneous hardening ("age-hardening") takes place, the tensile strength rising to about 60,000 pounds per square inch and the hardness to about 100. The elongation does not change much during the aging process, but there is a marked decrease in plasticity and workability.

It is to be noted that there is in the heat treatment of duralumin no "critical temperature" above which the material must be heated to produce hardening on quenching. It was discovered by Wilm and independently by E. Blough that the amount of hardening resulting from heat treatment increases progressively as the quenching temperature increases.

If duralumin is immersed in liquid air immediately after quenching from the high temperature treatment, age-hardening does not take place. On the other hand, if the material is heated to slightly elevated temperatures, say 212 to 302 degrees Fahr. (100 to 150 degrees Cent.), age-hardening takes place more rapidly than at room temperature. At still higher temperatures, or even after very prolonged heating at 212 to 302 degrees Fahr. (100 to 150 degrees Cent.), the hardening is succeeded by softening.

After considering the possible effects of the various constituents occurring in complex alloys of the duralumin type, Merica concluded that the heat treatment effects were due to solution, precipitation and particle growth of $CuAl_2$, substantially as described above in connection with the binary aluminum-copper alloys. This explanation accounts readily for the fact that the amount of hard-

ening during aging increases as the temperature of quenching increases.

"At higher quenching temperatures more and more $CuAl_2$ is dissolved in solid solution. After quenching, the $CuAl_2$ is in excess of its solubility; the higher the quenching temperature the greater is the excess, and this is precipitated during aging. The hardening is in proportion to the amount of the highly dispersed $CuAl_2$ formed."[1]

This effect of quenching temperature on hardening holds, of course, only as long as there is enough copper in the alloy to form a saturated solid solution at the maximum temperature employed in heat treatment. Another aspect of the same principle is that, for a given quenching temperature, the potentiality for hardening increases with the copper content of the alloy until there is enough copper present to form a saturated solid solution.

In regard to the aging phenomena, it will again be of interest to quote from Merica.[1]

"Upon aging a quenched specimen at 392 degrees Fahr. (200 degrees Cent.), for example, the hardness first increases to a maximum and afterward decreases. During that aging there has been first a formation of fine nuclei of $CuAl_2$ followed by coalescence of these particles into ones of larger size. There is, therefore, a certain average size of particle of $CuAl_2$ for which the hardness of the material is a maximum; atomic dispersion of the solute, $CuAl_2$, is not the dispersion that produces the maximum hardness, but some intermediate one between it and that at which the particles become visible by ordinary means."

The degree of dispersion conducive to maximum hardness was termed "critical dispersion."

Later work has indicated that Merica's theory was correct in its general principles, although in the case of duralumin one or more other constituents, notably Mg_2Si, may be involved. Duralumin will be further discussed subsequently in this paper.

Slip Interference Theory

An attempt has been made[3] to analyze the mechanism of the hardening effect in duralumin so that the necessary conditions could be defined for all analogous cases. At first thought it seems a

[3] Jeffries, Zay, and Archer, R. S., Slip Interference Theory of the Hardening of Metals, *Chemical and Metallurgical Engineering*, 1921, Vol. 24, 1057.

little difficult to understand how a relatively small quantity of a constituent such as $CuAl_2$ scattered through grains of aluminum in the form of disconnected particles can greatly harden the aluminum. When the nature of hardness and the mechanism of deformation in metals are considered an explanation becomes apparent.

All of the specific properties which are described by the general term hardness may be considered as resistance to deformation. The deformation of metals ordinarily takes place by the movement of crystalline fragments along the planes of weakness called slip planes. It seemed that the hardening effect of disconnected particles of a hard constituent could be attributed to their action as "keys" preventing the motion of crystalline fragments along slip planes. It is easily seen that the strengthening effect of the hard particles increases with their number rather than with their size. Even in case all possible planes of slip are keyed at some points, more effective slip interference must result from diminishing the size and increasing the number of the particles. We thus have the rule that the hardening effect of a given amount of hard dispersed substance is greater the smaller the particles. This rule must reach a limit as the particles approach atomic dimensions. We do not know how small the crystal may be and still possess the characteristic hardness and strength of the substance.

Extending the idea further, it appeared that all of the known methods of hardening metals could be attributed to this general principle of slip interference. Hardening by the formation of a finely divided precipitate of a hard constituent is one special case of this general theory.

The fact that disconnected particles of a constituent scattered through a metal could produce a certain amount of hardening was of course not previously unknown. It was recognized, for example, that tool steel containing cementite entirely in the spheroidized condition is much harder and stronger than pure iron. It was a rather prevalent idea, however, that in such cases it was necessary for the dispersed constituent to have considerable adhesion for the metal in which it occurred. The slip interference idea indicated that hardening should result from the presence of particles of any constituent materially harder, and hence stronger, than the metal in which it is embedded. Hardening would therefore be expected

to result from the presence of particles of non-metallic inclusions. On consideration of existing evidence it was recognized that copper is hardened by the presence of copper oxide and tungsten by the presence of thoria. In these cases there is probably very little specific attraction between the dispersed particles and the matrix.

Analysis of Heat Treatment for Hardening

Dispersions of sufficient fineness to produce a high degree of hardness are as a rule produced only by the decomposition of a supersaturated solid solution. The hardening process thus involves first the production of a supersaturated solid solution, and secondly the decomposition of this solution under such conditions that the particles of precipitate formed are very small and remain small.

As a rule the solubility of hard constituents such as intermetallic compounds in metals increases with increasing temperatures. Usually therefore the first step in the process consists in subjecting an alloy of suitable composition to heat treatment at a relatively high temperature with the object of bringing as much as possible of the hard constituent into solid solution. This heat treatment may be referred to as a "solution heat treatment."

A certain amount of time is of course required to bring about diffusion and solution. It is therefore necessary not only to heat the alloy to a suitable temperature, but also to hold it at temperature for a sufficient length of time to effect the desired solution. The length of time required in general increases with the distances through which diffusion must take place. Solution can be brought about more quickly in wrought materials than in cast materials, and in cast materials the process is more rapid the smaller the grain size.

It is usually necessary to avoid any fusion of the material during the solution heat treatment. In the case of the aluminum-copper alloys, for example, the temperature of the solution heat treatment should lie below the line ABE. As a matter of fact it is usually desirable to stay below the eutectic temperature BE since on passing this temperature fusion will occur around any undissolved particles of $CuAl_2$

The retention of a supersaturated solid solution at temperatures below its region of stability is in general favored by rapid

cooling.[4] It may not always be possible to retain such a solution without decomposition even by the most rapid means of cooling available. The retention of the solid solution is also usually favored by superheating above the temperature required to effect complete solution. This is generally attributed to the more complete destruction of crystalline nuclei. The slowest velocity of cooling which will prevent the decomposition of the solid solution may be called the critical cooling velocity. This may differ greatly in various cases.

Once the solid solution has been cooled to a temperature below that indicated by the solubility curve there is a tendency for precipitation to take place. This tendency becomes stronger the greater the degree of undercooling, but is opposed by an ever increasing rigidity as the temperature becomes lower. Thus in the case of duralumin it is found that at the temperature of liquid air atomic mobility is so slight that precipitation does not take place at all. At room temperature the process takes place slowly and to such a limited extent that no substantial decrease in hardness has been observed even after aging for many years. At temperatures of 212 to 392 degrees Fahr. (100 to 200 degrees Cent.) precipitation is much more rapid and after certain periods of time, which decrease as the temperature rises, the precipitated particles grow to such size that the initial age-hardening is succeeded by softening.

The temperatures suitable for the precipitation process vary widely according to the alloy involved. Usually these temperatures will lie not far below the temperatures required to cause annealing in the same alloys in the cold worked condition. If the optimum temperature for precipitation hardening happens to lie below room temperature, then the alloys will unavoidably "over-age" or soften on standing for sufficient time at room temperature. Such a condition may be looked for in alloys consisting chiefly of metals which recrystallize at or below room temperature, notably lead, tin, and zinc.

The time required for the precipitation process is greater the lower the temperature at which it is carried out. In many cases it has been found that prolonged heating at a low temperature gives

[4] It has been shown that the retention of austenite in steel may be favored by an intermediate rate of cooling as by quenching in oil instead of water. In this case, however, the conditions are somewhat different from those here under discussion.

greater hardness than short heating at a higher temperature.*

Aluminum-Magnesium-Silicon Alloys

It was shown by Wilm that alloys made by the addition of magnesium to commercial aluminum could be hardened by quenching from around 932 degrees Fahr. (500 degrees Cent.) followed by aging at room temperature. Later work at the National Physical Laboratory, Teddington, England, indicates that the heat treatment effects are in this case due to solution and precipitation of a compound Mg_2Si formed by the magnesium with the silicon present as an impurity in commercial aluminum. It was also shown that the solubility of this compound in solid aluminum decreases with decreasing temperature very much as does that of $CuAl_2$. After experimenting with alloys of various compositions and with various heat treatments it was concluded that the age-hardening effect was proportional to the amount of the compound Mg_2Si dissolved in the solution heat treatment. The age-hardening is presumably due to the formation of very small highly dispersed particles of Mg_2Si.

New Develpments in High Strength Aluminum Alloys

The theory of hardening which has been described above played a considerable part in the development of some new aluminum alloys of high strength.[5] Alloys of the duralumin type are rather difficult to work and their commercial development has been somewhat retarded by this characteristic. In view of the potentially great utility of strong aluminum alloys, it seemed desirable to find means of producing such alloys with better fabricating properties, and work was undertaken with this object in the laboratories of the Aluminum Company of America several years ago. This work has resulted in the production of two new types of aluminum alloy of the strong alloy class having distinctly improved fabricating

*Some as yet unpublished results obtained by P. V. Faragher in the laboratories of the Aluminum Company of America verify this conclusion for aluminum-copper alloys both with and without magnesium. This work shows that on aging at 100 degrees Cent. the hardness does not reach its maximum even after three months, but that if the aging is carried out at 125 degrees Cent. for about eight to twelve days or at 150 degrees Cent. for one or two days, the tensile strength passes through a maximum and then decreases. The maximum strength obtainable at 150 degrees Cent. was, for the alloys examined, about 2,000 to 3,000 pounds lower than that obtainable at 125 degrees Cent. A similar difference was found between the latter value and that obtained after aging at 100 degrees Cent. for three months.

[5]Archer, R. S., and Jeffries, Zay, New Developments in High Strength Aluminum Alloys, Paper of American Institute of Mining and Metallurgical Engineers, No. 1415-E, issued with *Mining and Metallurgy*, February, 1925

qualities. The knowledge gained has also made it possible to produce alloys of the duralumin type with greater strength and hardness than previously obtained.

One of the new types of alloy referred to consists chiefly of aluminum and copper. The binary aluminum-copper alloys especially when made from metal of ordinary purity show very little age-hardening at room temperature after the solution heat treatment. They are, however, susceptible to marked hardening by reheating at 212 to 347 degrees Fahr. (100 to 175 degrees Cent.). By proper selection of alloy composition and conditions of heat treatment it has been found possible to produce a material having approximately the same physical properties as duralumin and fabricating qualities superior in many respects to those of duralumin.

The other type of alloy referred to consists chiefly of aluminum, magnesium and silicon. As mentioned above, these alloys show distinct age-hardening at room temperature after the solution heat treatment. The strength attained under these conditions, however, is not very great—usually well under 40,000 pounds per square inch. The Brinell hardness is only about 70 which is not enough for satisfactory machinability nor in many cases for satisfactory resistance to deformation. The yield point is also comparatively low. In the laboratories of the Aluminum Company of America it was found that the hardness, strength and elastic ratio of these alloys could be further increased to a marked extent by artificial aging at about 212 to 347 degrees Fahr. (100 to 175 degrees Cent.). By this means it was possible to obtain a tensile strength as high as 54,000 pounds per square inch, a Brinell hardness of 120, and a yield point of about 40,000 pounds per square inch. These changes were of course accompanied by a decided decrease in plasticity.

This discovery greatly increased the potential usefulness of this type of alloy. It was also found that these alloys possess excellent working qualities, the capacity for cold working in particular being quite remarkable. This may be judged from the fact that in the annealed condition the tensile strength may be as low as 15,000 pounds per square inch and the Brinell hardness less than 30, values only slightly higher than those for commercially pure aluminum.

It will be apparent that duralumin may be regarded as a combination of the two types of alloy here described, inasmuch

as it contains copper and magnesium, both intentionally added, together with some silicon which was present in the aluminum ingot as an impurity. The amount of silicon in aluminum ingot is quite variable, however; hence the amount obtained in duralumin by accident may or may not be most suitable. In view of the beneficial effects of silicon, alloys have been prepared of the duralumin type, but containing more than the usual amount of silicon. One example of this type of alloy contained 4.32 per cent copper, 0.78 per cent silicon, 0.50 per cent magnesium, and 0.49 per cent manganese. After forging, this alloy was heated to a temperature between 932 and 960 degrees Fahr. (500 and 515 degrees Cent.) and quenched in water. After aging for three days at room temperature the tensile strength was 59,300 pounds per square inch and the elongation 25 per cent. Artificial aging for 70 hours at 248 degrees Fahr. (120 degrees Cent.) gave a strength of 66,900 pounds per square inch with an elongation of 20.5 per cent, while 20 hours at 302 degrees Fahr. (150 degrees Cent.) gave a tensile strength of 75,300 pounds per square inch and an elongation of 13 per cent. Thus by combining the new information regarding composition and heat treatment it has been possible to produce an alloy of the duralumin type having greater strength and hardness than hitherto produced by heat treatment.

Magnesium-Aluminum Alloys

The work on aluminum alloys which has just been described led to the development of a process for hardening magnesium-base alloys, especially the magnesium-aluminum alloys.[6] For example, a chill casting of an alloy containing 12 per cent aluminum was found to have a Brinell hardness of 71 as cast. After a solution treatment of 15 hours at 788 degrees Fahr. (420 degrees Cent.) and quenching in water, the hardness was 66. The casting was then given a precipitation treatment of 40 hours at 302 degrees Fahr. (150 degrees Cent.) after which the hardness was 111.

In the magnesium-aluminum system a eutectic occurs at about 32 per cent aluminum and 815 degrees Fahr. (435 degrees Cent.). The eutectic mixture consists of a magnesium-aluminum solid solution and a hard constituent reported to be Mg_3Al_2. The solubility of aluminum in solid magnesium is about 12 per cent at the

[6] Jeffries, Zay, and Archer, Robert S., U. S. Patent 1,592,302.

eutectic temperature, and decreases with decreasing temperature.[7]

In the most recent published investigation,[7] the solubility of aluminum in magnesium is reported to be about 9 per cent at 392 degrees Fahr. (200 degrees Cent.) and perhaps 8 per cent at room temperature. On this basis it would be expected that alloys containing less than 8 per cent aluminum would not show the age-hardening phenomenon. As a matter of fact marked hardening was found in alloys containing 8 per cent or more of aluminum, and distinct hardening (from 51 to 59 Brinell) was observed in an alloy containing 6 per cent aluminum. Assuming the validity of the precipitation theory of hardening, this indicates that the solubility of aluminum in solid magnesium at room temperature is less than 6 per cent.

The detection of this discrepancy suggests that hardening experiments may profitably be used to check microscopic examination in the study of this type of alloy constitution.[8] In fact, the hardness changes may sometimes offer the most sensitive test of solid solubility.

The hardening phenomena in the magnesium-aluminum alloys exhibit some other features of interest to which attention may be called.

Alloy Composition versus Extent of Hardening.—The increase in hardness effected by the precipitation treatment was found to be greatest with an aluminum content of about 12 per cent, that is, at about the maximum solubility of aluminum in solid magnesium.

Alloy Composition versus Rate of Hardening.—It was found that the *rate* as well as the extent of precipitation hardening increased with the aluminum content of the alloy up to about 12 per cent aluminum, decreasing somewhat with 14 and 16 per cent aluminum. In one experiment, for example, specimens of alloys containing 6, 7, 8, 10, 12, 14 and 16 per cent aluminum were heated 15 hours at 779-788 degrees Fahr. (415-420 degrees Cent.), quenched in water, and reheated at 392 degrees Fahr. (200 degrees Cent.). The alloys containing upward of 10 per cent aluminum attained maximum hardness at the end of about 7 hours, after which they softened slightly. In the case of the alloys containing 6, 7, and

[7]Hanson, D., and Gayler, M. L. V., The Constitution of the Alloys of Aluminum and Magnesium, *Journal*, Institute of Metals, 1920, No. 2, p. 201.

[8]Science of Metals, Jeffries and Archer, McGraw-Hill Book Co., pp. 329-330.

8 per cent aluminum, the hardness was still increasing after about 60 hours.

Initial Softening.—In the early stages of the precipitation treatment, the alloys of lower aluminum content sometimes showed slight softening.

Cooling Rate versus Hardening Rate.—The rate of cooling from the solution heat treatment was observed to have a marked effect on the rate of precipitation hardening. For example, chill cast specimens of an alloy containing 12 per cent aluminum were heated 18 hours at 779-788 degrees Fahr. (415-420 degrees Cent.) after which some were quenched in water while others were air cooled. The quenched material had a Brinell hardness of 74 while the air cooled material was harder, (83 Brinell) due presumably to some precipitation effect during cooling. On reheating at 302 degrees Fahr. (150 degrees Cent.), the quenched material attained its maximum hardness of 114 in 20 hours, when the air cooled material showed a hardness of only 92, which increased to 111 after 64 hours and 114 after 108 hours. In this case the different rates of cooling from the solution treatment caused marked differences in the rate of hardening, but not in the maximum hardness attained.

Precipitation Temperature versus Hardness.—It was also found that the maximum hardness that could be attained was in general higher the lower the temperature of the artificial aging or precipitation heat treatment. The time required to reach this maximum hardness was, however, longer the lower the temperature. Similar relations were found in the aluminum-base alloys and it seems that they are generally true in cases of precipitation hardening. The supersaturated solid solution must be allowed to stand or age at a temperature sufficiently high to permit precipitation to take place. Quenched duralumin for example does not undergo age-hardening in liquid air. As the temperature of aging is raised the extent of the hardening appears to increase up to a certain temperature. Higher temperatures produce more rapid aging but lower final hardness. This is presumably due to the fact that at these higher temperatures the precipitated particles grow so rapidly that softening is taking place simultaneously with hardening, so that the maximum hardness is never attained.

Copper-Iron Alloys

In the course of an investigation of the effect of small quantities of iron on the properties of copper at the National Physical Laboratory, Hanson and Ford[9] found that solid copper dissolves about 4 per cent of iron at 2012 degrees Fahr. (1100 degrees Cent.), the solubility falling to less than 0.2 per cent at 1382 degrees Fahr. (750 degrees Cent.), as shown in the diagram in Fig. 2. In view

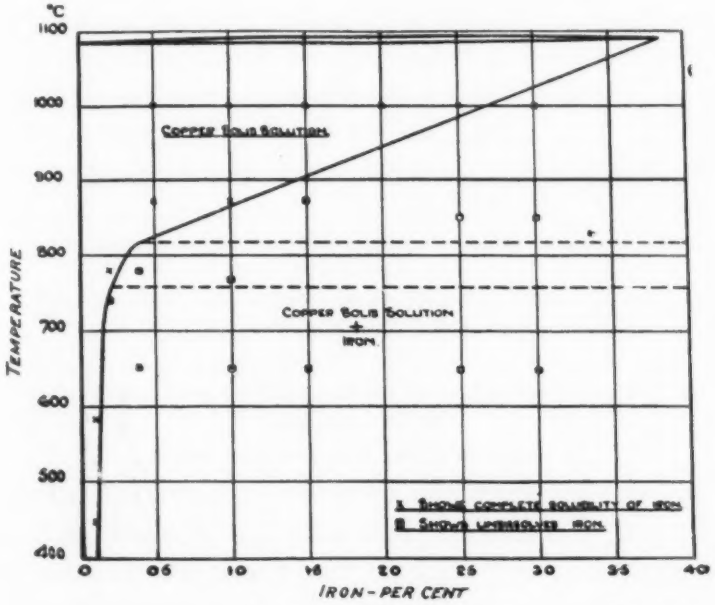


Fig. 2—Constitution of Copper-Iron Alloys (Hanson and Ford)

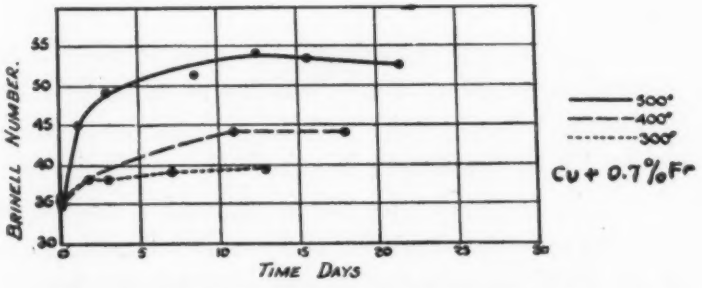


Fig. 3—Precipitation Hardening in Copper-Iron Alloy (Hanson and Ford)

of the large difference in the solubility of iron in solid copper at high and low temperatures it was expected that considerable variaiton in physical properties could be produced by heat treatment.

[9]Hanson, D., and Ford, Grace W., Investigation of the Effects of Impurities on Copper. Part II.—The Effect of Iron on Copper, *Journal*, Institute of Metals, Vol. XXXII, No. 2, 1924.

Fig. 3 shows the precipitation hardening of an alloy containing 0.7 per cent iron, quenched from 2012 degrees Fahr. (1000 degrees Cent.) and reheated at 572, 752 and 932 degrees Fahr. (300, 400 and 500 degrees Cent.). No appreciable age-hardening could be detected at room temperature. Tensile tests were also made but it was found that the increases in tensile strength were not as great

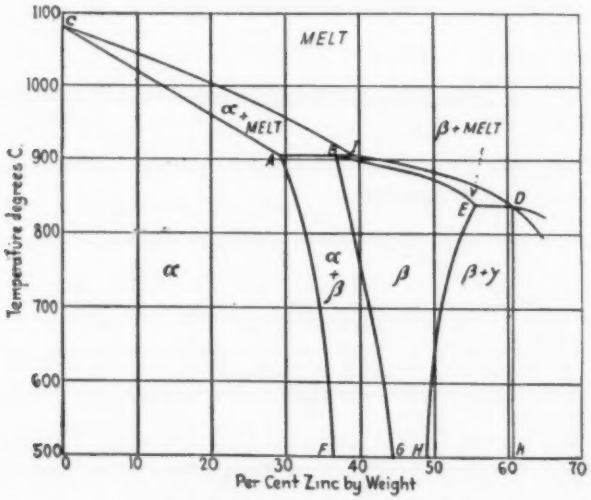


Fig. 4—Constitution of Copper-Zinc Alloys

as would be expected from the increases in Brinell hardness.

Hanson and Ford were not able to produce substantially greater hardening by low temperature precipitation treatments of long duration than by treatments of short duration at higher temperatures. This constitutes an exception to the general rule stated above.

In the cases previously considered the precipitated constituents have been intermetallic compounds which are relatively very hard and strong. Commenting on the results of their work the authors[9] say 'The effect of heat treatment is relatively small, in spite of the large differences in solubility of the iron, and a considerable improvement in strength as a result of heat treatment, such as can be obtained in steel and some aluminum alloys, has not been realized. The relatively small effect of the 'iron' constituent on the strength is possibly due to the fact that this constituent is essentially metallic iron (containing some dissolved copper), and does not itself possess the great strength and hardness of intermetallic compounds, which usually have a more considerable effect on the mechanical properties.''

Alpha-Beta Brass

Although the iron constituent precipitated in the copper-iron alloys just described is soft as compared with some of the intermetallic compounds, it is hard as compared with the copper in which it occurs. A case of considerable interest has been investigated in which hardening results from the precipitation of a constituent which by itself is softer than the matrix in which it is precipitated.

The diagram in Fig. 4 shows the constitution of the copper-zinc alloys rich in copper. From this diagram it will be seen that alloys containing from 38 to 44 per cent zinc consist entirely of the beta constituent at temperatures above the line BG. On cooling slowly below this line the alpha solid solution precipitates. By heating above the line BG and cooling rapidly the beta constituent will be retained and on reheating at suitable temperatures the alpha

Table I

Effect on Hardness of Reheating Quenched Samples of Alloys A and B at 392 degrees Fahr. (200 degrees Cent.)

Alloy	Brinell Hardness Numbers Prior to Reheating	Brinell Hardness Numbers Subsequent to Reheating	Time of Reheating Hours	Percentage Increase in Hardness	Structure Revealed by Microscope
A	130	130	¼	nil	
Cu 56.66	133	133	½	nil	
Zn Bal.	133	133	1	nil	
	130	130	2	nil	Beta
Quenched	133	133	4	nil	
from	136	138	8	1.5	
720° C.	130	140	12	7.7	
	133	150	16	12.8	
	130	143	40	10.0	
B	146	146	¼	nil	
Cu 52.02	150	150	½	nil	
Ni 3.86	150	150	1	nil	
Zn Bal.	150	150	2	nil	
	150	150	4	nil	Beta
Quenched	146	146	8	nil	
from	146	152	12	4.1	
770° C.	146	158	16	8.2	
	150	158	40	5.3	

constituent may be caused to precipitate in a finely divided form. For their heat treatment experiments Ellis and Schemnitz[10] selected an alloy (A) containing 56.66 per cent copper, remainder zinc, and an alloy (B) containing 52.02 per cent copper, 3.85 per cent nickel, remainder zinc. The nickel functions in a similar man-

[10] Ellis, O. W., and Schemnitz, D. A., Experiments on the Heat Treatment of Alpha-Beta Brass, paper of the American Institute of Mining and Metallurgical Engineers, No. 1348-N, July, 1924.

ner to the copper so that alloy B may be regarded as analogous to a copper-zinc alloy of somewhat higher copper content.

By preliminary experiments it was found that suitable temperatures for the solution heat treatment of these two alloys were 1328 and 1418 degrees Fahr. (720 and 770 degrees Cent.) respectively. After quenching from these temperatures samples were reheated at various temperatures, Brinell hardness determinations having been made before reheating. The results of reheating quenched samples at 392 degrees Fahr. (200 degrees Cent.) are shown in Table I. A maximum increase of hardness of 12.8 per cent was observed. Definite age-hardening seemed to occur even at room temperature; several samples showed average increase of hardness of 9.4 per cent.

A copper-zinc alloy consisting entirely of the alpha solution is softer than one consisting of the beta constituent. Ellis and Schemnitz considered it difficult to imagine how soft particles of alpha constituent could act as "keys" to prevent slip in the beta constituent. It would appear that this difficulty is sufficiently answered by pointing out that in order for a precipitated particle to act as an effective key it is only necessary that its resistance to deformation in a random direction be greater than the resistance to deformation along a slip plane of the matrix in which it is embedded. Although beta brass is on the whole harder than alpha brass, still it is quite reasonable to suppose that the strength of alpha brass in directions which do not correspond to its planes of easy slip is considerably greater than that of beta brass along its planes of slip.

It may be mentioned that the hardening produced by the precipitation of the alpha constituent was not accompanied by increased strength. As a matter of fact the authors say that the precipitation of alpha has a deleterious influence on tensile properties which they attribute to the precipitation of some of the alpha at the grain boundaries. In this location the presence of the weaker alpha constituent lowers the strength of the brass as a whole.

Lead-Antimony Alloys

The lead-antimony system has for a long time been considered a typical example of a simple eutectiferous system with substantially no mutual solid solubility of the components. It has recently

been shown[11], however, that antimony dissolves in solid lead up to 2.45 per cent at the eutectic temperature 477 degrees Fahr. (247 degrees Cent.), the solubility decreasing to less than 0.5 per cent at room temperature as shown in the constitution diagram in Fig. 5. A marked age-hardening phenomenon was discovered in lead-antimony alloys containing more than 0.5 per cent antimony

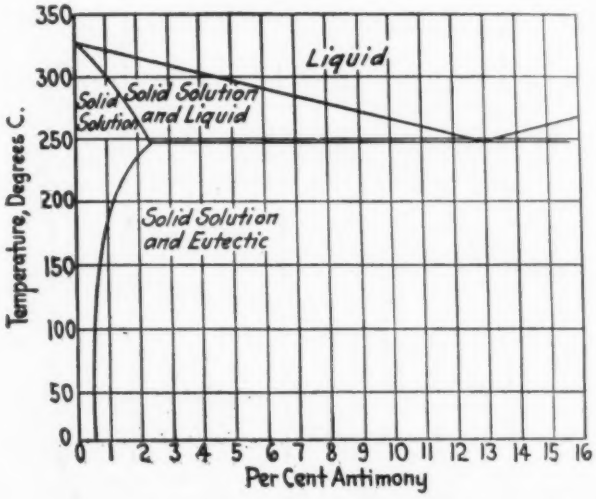


Fig. 5—Constitution of Lead-Antimony Alloys (Dean, Zickrick and Nix)

after quenching from above 392 degrees Fahr. (200 degrees Cent.), as illustrated in the diagram in Fig. 6. This hardening is supposedly due to the precipitation of nearly pure antimony which is quite hard as compared with lead.

The investigation of this system brought out confirmation of some of the general characteristics of precipitation hardening described above. It was found, for example, that the maximum hardening effect occurred at about 2.5 per cent antimony. "This fact may be explained on the dispersion theory by assuming that in alloys containing more than 2.5 per cent antimony, some relatively large primary crystals of the separating constituent will be present after quenching, which will serve as nuclei and prevent as complete a dispersion of the separating material as in the alloy which represents the maximum solid solubility of the series."[12]

[11]Dean, R. S., Zickrick, L., and Nix, F. C., The Lead Antimony System and Hardening of Lead Alloys, Paper No. 1539-E, American Institute of Mining and Metallurgical Engineers, February, 1926.

[12]Dean, R. S., Hudson, W. E., and Fogler, M. F., The Lead Antimony System II, *Industrial and Engineering Chemistry*, 1925, Vol. 17, 246.

It was found that "heating beyond the time necessary to get all the antimony into solution (in the solution heat treatment) has practically no effect on the tensile strength but increases brittleness. This is probably due to grain growth of the lead matrix."[11]

To get hardening effects near the maximum, it was found neces-

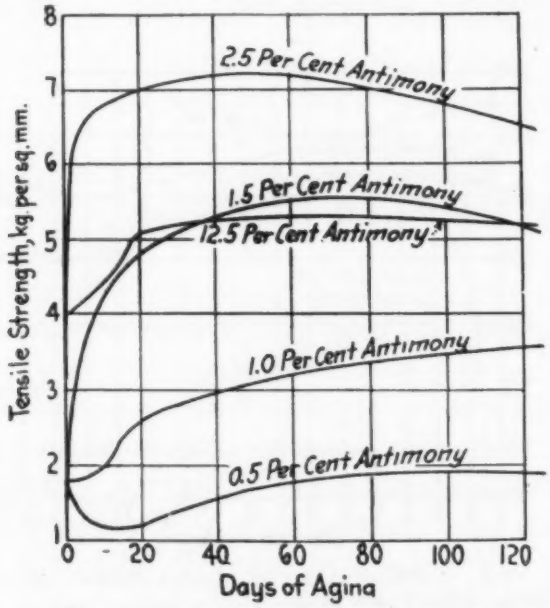


Fig. 6—Effect of Room Temperature Aging on Strength of Lead-Antimony Alloy Wires Quenched from 238-240 degrees Cent. (Dean, Zickrick and Nix)

sary to heat above 446 degrees Fahr. (230 degrees Cent.) for the solution treatment.

The effect of rate of cooling from the solution treatment was investigated in the case of a 2.5 per cent antimony alloy heated at 455 degrees Fahr. (235 degrees Cent.) and cooled by various methods. It was found, as in the case of the magnesium alloy containing 12 per cent aluminum, that the rate of precipitation hardening increased with the rate of cooling from the solution treatment. The maximum hardness attained after aging was less the slower the cooling.

The lead-antimony case is of especial interest in view of the fact that the matrix, lead, anneals at or below room temperature. It would be expected, as pointed out above, that the quenched alloys would "over-age," that is, begin to soften, on prolonged standing at room temperature. This was found to be true, all of the alloys showing softening after attaining maximum hardness

in 10 to 16 days, following a solution treatment of 15 minutes at 460 degrees Fahr. (238 degrees Cent.) and quenching in cold water.

Softening at elevated temperatures was studied by reheating 2.5 per cent antimony alloy wire which had been heat treated and aged 10 days at room temperature. At each temperature (77, 100, 125 and 140 degrees Cent.) the tensile strength fell rapidly at first and then approached what was considered an equilibrium

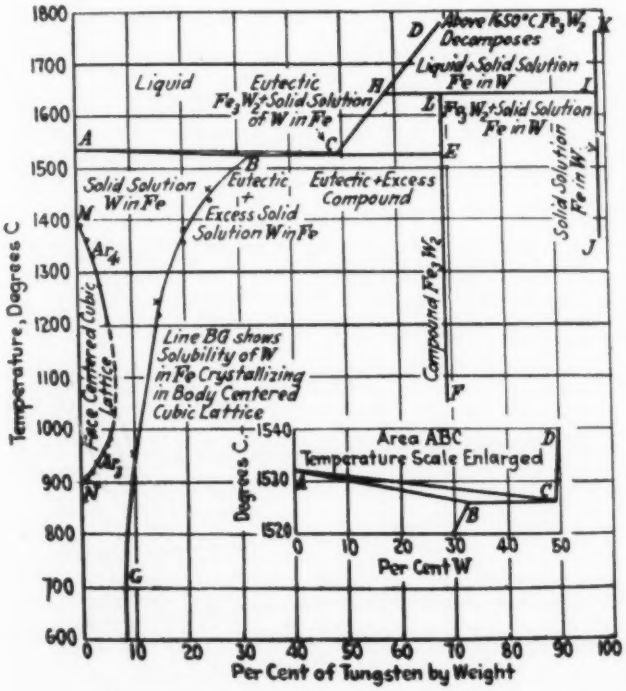

Liquid
Eutectic
Fe3W2+Solid Solution of W in Fe
Above 1550°C Fe3W2 Decomposes
Liquid+Solid Solution Fe in W
Fe3W2+Solid Solution Fe in W
Solid Solution W in Fe
Eutectic + Excess Solid Solution W in Fe
Eutectic+Excess Compound
Solid Solution Fe in W
Compound Fe3W2
Ar4
Face Centered Cubic Lattice
Line BG shows Solubility of W in Fe Crystallizing in Body Centered Cubic Lattice
Ar3
Area ABC Temperature Scale Enlarged
Degrees C.
Per Cent W
Temperature, Degrees C
Per Cent of Tungsten by Weight

Fig. 7—Constitution of Iron-Tungsten Alloys (Sykes)

value, lower the higher the temperature of the reheating. Extrapolating to 77 degrees Fahr. (25 degrees Cent.), it was concluded that the equilibrium or stable strength was about 78 per cent of the maximum noted strength and about twice that of the untreated wire.

Iron-Tungsten Alloys

In a recent investigation of the iron-tungsten system W. P. Sykes[13] found that at 2777 degrees Fahr. (1525 degrees Cent.) iron will hold in solid solution about 33 per cent tungsten. The solubility decreases to less than 10 per cent at 1112 degrees Fahr. (600

[13]Sykes, W. P., The Iron-Tungsten System, American Institute of Mining and Metallurgical Engineers, Paper No. 1515-C, issued with *Mining and Metallurgy*, February, 1926.

degrees Cent.) as shown in the diagram in Fig. 7. When the amount of tungsten present exceeds the solubility limit a hard and brittle constituent appears which is believed to be a compound

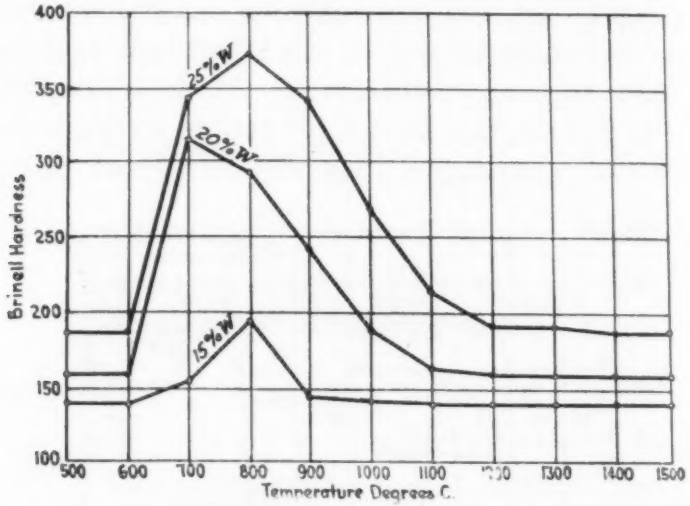


Fig. 8—Brinell Hardness of Iron-Tungsten Alloys Water-Quenched from 1500 degrees Cent. and heated 1 hour at temperatures indicated. (Sykes)

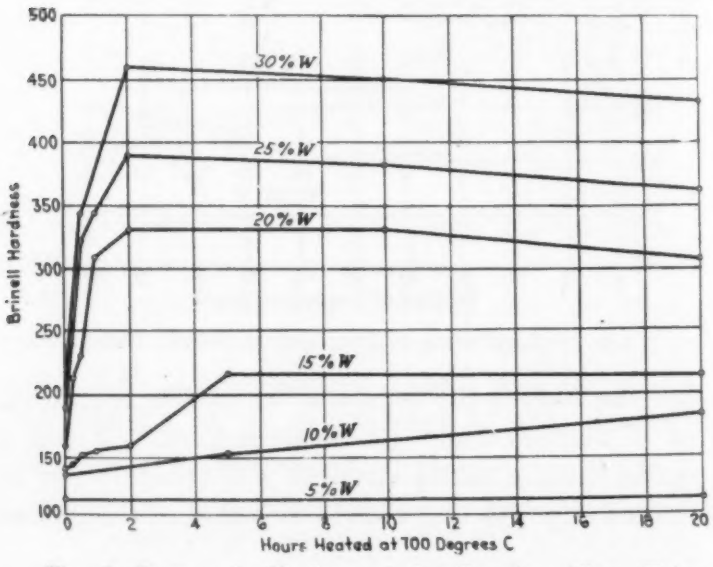


Fig. 9—Increase in Hardness of Iron-Tungsten Alloys Water-Quenched from 1500 degrees Cent. and heated at about 700 degrees Cent. (Sykes)

Fe_3W_2. It was found possible to produce hardening by heat treating to obtain a large amount of tungsten in solid solution, cooling this solid solution rapidly and reheating at suitable temperatures, as is illustrated by the results shown in Figs. 8 and 9.

Referring to the iron-tungsten diagram it will be noted that

the allotropic change in iron which plays an all important part in the hardening of ordinary steel is entirely absent in the iron-tungsten alloys under consideration. The presence of tungsten in iron lowers the A_4 point and raises the A_3 point. With tungsten contents of 6 per cent or more the face-centered cubic allotrope gamma iron does not occur. The alloys under consideration cool from 2777 degrees Fahr. (1525 degrees Cent.) to room temperature without any allotropic change in the iron. The marked hardening is presumably due to the precipitation of the hard constituent Fe_3W_2 in finely divided form.

Because of the high temperature required to produce precipitation and particle growth, the hardness of the heat treated iron-tungsten alloys is comparatively stable at temperatures up to 1112 degrees Fahr. (600 degrees Cent.). This is the property called red hardness in connection with high speed steel and means the ability to retain hardness at temperatures above red heat. Precipitation hardening similar to that just described for iron-tungsten alloys has been found in the case of iron-molybdenum alloys and the resistance to softening at elevated temperatures, or in other words, the red hardness of some of these alloys is superior to that of high speed steel. In certain applications an alloy of this type has given as much as seventy times the length of service of the best high speed steel.

In addition to the fact that the allotropic change in iron is not involved in this hardening process, it is to be noted that these alloys are entirely free from carbon. Whereas the hardness of hardened carbon tool steel is to be attributed chiefly to the extremely fine grain which results from the retarded allotropic change and to the presence of carbon both in solution and in the form of very small particles of cementite, the hardening of the new iron-tungsten and iron-molybdenum alloys is due entirely to the precipitation of a hard constituent throughout fairly large grains.

Cause of Age-Hardening in Duralumin

Duralumin is a complex alloy containing aluminum, copper, magnesium and silicon. The age-hardening of aluminum-magnesium-silicon alloys, due presumably to precipitation of Mg_2Si, has been described above. Since alloys of this type exhibit marked hardening on aging at room temperature, whereas the binary

aluminum-copper alloys show relatively little age-hardening at room temperature, it is a rather natural inference that the room temperature aging of duralumin is due to the precipitation of Mg_2Si rather than of $CuAl_2$.

This conclusion was reached by the investigators at the National Physical Laboratory, [14, 15, 16] and has been supported by others.

There is some new information which suggests very strongly that the room temperature age-hardening of duralumin may be due to something other than the precipitation of Mg_2Si. The recent production of aluminum of very high purity by an electrolytic refining process[17] has made it possible to prepare alloys containing copper and magnesium but almost free from silicon and hence from Mg_2Si. A wrought alloy of this type containing 3.55 per cent copper, 0.48 per cent magnesium, 0.03 per cent iron, and 0.02 per cent silicon was quenched from 990 degrees Fahr. (533 degrees Cent.). The tensile strength immediately after quenching was approximately 36,000 pounds per square inch, which increased on aging at room temperature to about 53,000 pounds per square inch. The corresponding increase in Brinell hardness was from 78 to 107 (10 millimeter ball-1000 kilograms). Another alloy, of about the same degree of purity, containing about 4.0 per cent copper and 0.5 per cent magnesium was made into chill castings which were given a prolonged heat treatment and quenched from 1022 degrees Fahr. (550 degrees Cent.). The hardness was then 73 and increased to 112 after aging 25 days at room temperature. It will be noted that these age-hardening effects are fully as great as in ordinary duralumin containing substantial quantities of silicon.

Binary aluminum-copper alloys made from ordinary aluminum show only slight age-hardening at room temperature. In alloys

[14]Gayler, M. L. V., The Constitution and Age-Hardening of Alloys of Aluminum with Copper, Magnesium and Silicon in the Solid State, *Journal*, Institute of Metals, Vol. 28, 1922, pp. 213-244.

[15]Gayler, M. L. V., The Constitution and Age-Hardening of the Ternary Alloys of Aluminum with Magnesium and Copper, *Journal*, Institute of Metals, Vol. 29, 1923, pp. 507-526.

[16]Gayler, M. L. V., The Constitution and Age-Hardening of the Quaternary Alloys of Aluminum, Copper, Magnesium and Magnesium Silicide, *Journal*, Institute of Metals, Vol. 30, 1923, pp. 139-166.

[17]Frary, F. C., The Electrolytic Refining of Aluminum, *Transactions*, American Electrochemical Society, 1925.

made from aluminum of high purity, the extent of the age-hardening at room temperature is greater, but still not comparable with that when magnesium is present. The mechanism by which the presence of magnesium increases the age-hardening effect is not certain. There is little probability of precipitation of the compounds Mg_2Al_3, $MgCu_2$ or $CuMg_4Al_6$ in alloys containing around 4.0 per cent copper and 0.5 per cent magnesium. It seems more likely that the presence of the magnesium stimulates precipitation of $CuAl_2$. The atomic volume of magnesium (14.0) is considerably greater than that of aluminum (10.0). The presence of the magnesium in solid solution may therefore distend the space lattice of the aluminum enough to materially facilitate the diffusion which is necessary to the formation of a precipitate. This would be approximately equivalent to lowering the temperature necessary for precipitation.

Present Status of Precipitation Theory of Hardening

It is now generally recognized that metals can be hardened by dispersed constituents precipitated from solid solutions. This is a conception whose general validity and practical usefulness can scarcely be questioned. There are, of course, some questions regarding the details of the theory and its application in various specific cases.

Electrical Resistivity.—The chief objection to the precipitation theory as applied to duralumin is based on the changes in electrical resistivity which accompany age-hardening. It has been considered that in general the resistivity of an alloy is increased when an aggregate is changed into a solid solution, and conversely that the resistivity is decreased when a solid solution decomposes into an aggregate. Accordingly, the resistivity of an alloy would be expected to increase as the result of a solution heat treatment, and to decrease as the result of a precipitation treatment. In the lead-antimony alloys described above these changes occur with such consistency that the study of constitution and aging was based largely on measurements of resistivity. In duralumin, however, an apparent exception is encountered. The electrical resistivity is markedly increased by the solution treatment, as would be expected, but continues to increase slightly during room temperature aging. It has been considered by some that this increase in re-

sistivity is inconsistent with the idea of precipitation.

As a result of aging at elevated temperatures, the resistivity of duralumin and other heat treated aluminum alloys decreases. Fraenkel, for example, has suggested that a fundamental difference must exist between room temperature and "artificial" aging, and agrees with the precipitation theory only in regard to the latter.[18]

It has recently been observed [19] in some specimens of quenched duralumin that the initial increase in resistivity during room temperature aging was succeeded after prolonged aging by a decrease in resistivity. This indicates that room temperature aging is not fundamentally different from aging at elevated temperatures, as suggested by Fraenkel, but that the increased resistivity is merely characteristic of incomplete aging, although of a degree of aging accompanied by marked hardening.

The precipitation theory of age-hardening is itself a correction of the older idea that solid solutions possess maximum hardness. It seems quite logical that a similar correction may be necessary of the old idea that solid solutions necessarily possess maximum electrical resistivity.

It is not difficult to imagine possible causes for an increase in resistivity in some cases during the first stages of precipitation. Obviously the form and distribution of the precipitate are of importance. It has been found, as will be described below, that $CuAl_2$ tends to precipitate along grain boundaries. Perhaps this precipitate first forms in rather extended films or plates which subsequently coalesce. Another factor may be imperfect contacts, simulating voids, between the matrix and the newly formed particles. Inasmuch as the formation of a compound such as $CuAl_2$ involves the withdrawal of two atoms of aluminum from the matrix for each atom of copper, it is quite conceivable that the spaces occupied by these aluminum atoms in the space-lattice are not immediately filled by other aluminum atoms. This would appear to be especially likely at low temperatures of precipitaton, and in aluminum alloys it seems that the initial rise in resistivitity is less marked the higher the temperature of aging.

[18] Some of the work on electrical resistivity is discussed by K. L. Meissner, "Aging of Aluminum Alloys," *Metal Industry* (London), Vol. 38, 363, April 16, 1926.

[19] Taylor, C. S., Research Bureau, Aluminum Company of America, private communication.

Direct Evidence of Precipitation.—When Merica first proposed his theory, there was no visual evidence of precipitation during aging. He was unable to observe under the microscope the particles thought to have formed during room temperature aging, or even after reheating at 572 degrees Fahr. (300 degrees Cent.). Precipitated particles of $CuAl_2$ formed from aluminum-copper solid solutions by reheating at 572 degrees Fahr. (300 degrees Cent.) have since been observed and distinct microscopic evidence of precipitation has been found in specimens heated at 392 degrees Fahr. (200 degrees Cent.)[2] Particles precipitated during aging have also been observed in the lead-antimony alloys. [11]

Form and Distribution of Precipitated Particles.—In some discussion of the precipitation theory it has been considered for the sake of simplicity that the precipitated particles were spherical in shape, of equal size, and of uniform distribution. It is of course highly improbable that any of these conditions is ever fully realized. In the case of aluminum-copper solutions reheated around 572 degrees Fahr. (300 degrees Cent.), for example, it has been noted that precipitation and particle growth take place preferentially at the grain boundaries, and that the particles within the grains are shaped like rods rather than spheres. These rods seem to lie in certain directions within any one grain forming patterns suggestive of martensitic markings. Aluminum has the same crystal structure as gamma iron (face-centered cubic). These patterns suggest that the precipitate tends to form along the 111 planes as well as along the grain boundaries.

Mechanism of Hardening.—The general principle that metals may be hardened by highly dispersed constituents formed on the decomposition of supersaturated solid solutions has already proved useful in leading to the discovery of new hardening effects and the explanation of known phnomena. The usefulness of the general principle would be further increased if the mechanism by which the dispersed particles cause hardness were thoroughly understood. It was with this hope that the idea was advanced[3] that the disconnected particles act as keys increasing the resistance to motion along the slip planes of the crystalline matrix. This idea led directly to the conclusions that to produce hardening the dispersed particles should be harder than the matrix and that the dispersed constituent need not possess great specific attraction (adhesion) for

the matrix. These conclusions are consistent with subsequent developments. Logical consequences of the acceptance of these conclusions would be the concentration of effort to produce hardening on alloys in which *hard* constituents can be produced in a highly dispersed condition and the direction of attention to possible hardening effects of "foreign" particles such as oxides, which were formerly considered to be sources of weakness rather than possible sources of strength.

There have been objections to the "key" idea. The discovery of slight hardening by the precipitation of alpha in a matrix of beta brass has been thought by some to show that the dispersed particles need not be harder than the matrix, and hence that the "key" explanation could not be entirely correct. This has been considered above, where it was pointed out that it is only necessary that the dispersed particles be stronger in random directions than the matrix along its planes of easy slip.

Merica has called attention to the fact that a given quantity of hard substance dispersed in a crystal will expose the same shearing area on the slip planes regardless of the particle size. He draws the conclusion that a given quantity of hard constituent should have the same keying effect regardless of the particle size. In view of the fact that a given quantity of a hard constituent actually has a greater hardening effect when present in particles of a certain size, Merica suggests that the cause must be looked for in the higher surface tension forces of the small particles, which give them an effective sphere of influence beyond their true diameters.

In reply to this objection it has been pointed out[20] that actually the very small hard particles are seldom sheared through, but exert their hardening effect by diverting slip into more complex paths in the matrix.

Marked "hardening" effects are sometimes produced by small particles dispersed in a non-crystalline matrix, as by colloids in water. From this it has been argued by some that precipitation hardening in metals is merely a special case of a general effect of colloidal particles. This view however, would require that rather special chemical relations exist between the colloidal particles and the matrix. The stiffening of a liquid by colloidal particles is presumably due to a fixing of the molecules of the liquid by specific

[20]Science of Metals, Jeffries and Archer, McGraw-Hill Book Co., p. 400.

attractive forces to such an extent that the liquid takes on the characteristics of a solid. Conditions are entirely different in metals, where the known crystalline structure is recognized as a source of softness because of the existence of definite crystallographic plans of easy slip. It scarcely seems profitable to ignore this condition. The general colloid idea would mean that to produce hardening the dispersed particles must possess strong specific or "chemical" attraction for the matrix, which does not seem to be the case.

The structural conditions around small particles in a metal are of course not known with exactness. Regardless of the specific atomic forces involved there must be some distortion in the crystalline matrix immediately around the imbedded particle. This distorted region may be considered as enlarging the effective size of the "key," much as the distorted grain around a knot increases still further the difficulty of splitting a piece of wood through the knot.

Summary

Metals may be effectively hardened by highly dispersed particles within the grains, such as may be formed from supersaturated solid solutions. The typical process of hardening by this means consists in a solution heat treatment at a relatively high temperature followed by rapid cooling into a region of supersaturation, then by a precipitation treatment or aging to permit the formation of a very fine precipitate. The precipitate tends to grow through a critical particle size conducive to maximum hardness, beyond which the hardness decreases. Following are some tentative generalizations regarding the details of the subject:

1. Maximum concentration of the parent solid solution is favored by high temperature and prolonged heating in the solution heat treatment.

2. Maximum retention of the solid solution in a supersaturated condition is favored by rapid cooling to low temperatures.

3. At some sufficiently low temperature, different for different metals, the supersaturated solid solution is permanently stable.

4. At some temperature, usually in the neighborhood of the recrystallization temperature of the solvent metal, precipitation and particle growth take place so rapidly that maximum hardness is not attained, and that whatever hardness is attained at first subsequent-

ly decreases to a marked extent.

5. Some temperature intermediate between those described in (3) and (4) is usually conducive to the formation and retention of a fine precipitate, with consequently great hardening effects.

6. The normal course of the hardness changes during the precipitation treatment include:

a. An initial softening. Not always observed.
b. Rise of hardness to a maximum.
c. Decrease of hardness to a substantially stable value which is higher the lower the temperature of treatment. This decrease may not be observed, even after years, as in the room-temperature aging of duralumin.

7. The rate of precipitation-hardening increases with:

a. Temperature of the precipitation treatment.
b. Rate of cooling from solution heat treatment.
c. Concentration of the hardening element or elements in the alloy, at least up to the composition of the most concentrated solution that can be formed during the solution treatment.

8. Maximum hardening effect is favored by:

a. Increasing concentration of hardening elements up to maximum solid solubility. Greater amounts of the hardening elements sometimes cause the precipitation hardening effect to be less.
b. Low temperature and long time in the precipitation treatment. The temperature must of course be sufficiently above the region of low temperature stability. Possibly a useful criterion of this condition is the softening which follows maximum hardness as the precipitation treatment is prolonged. Failure to observe this softening may indicate that the temperature of the precipitation treatment is too low for maximum hardening.

9. The increased hardness caused by precipitation is accompanied by decreased plasticity.

10. Ductility, as measured by per cent elongation in the tensile test, may increase during the first stages of precipitation hardening but usually decreases by the time maximum hardness is attained.

11. A high combination of strength and ductility is favored

by low temperature and long time, in the precipitation treatment, rather than by high temperature and short time.

The Author

ROBERT S. ARCHER

R. S. Archer was born in 1895 and received his early education in the Detroit Public Schools. In 1916 he received the degree of Bachelor of Chemical Engineering and in 1917 the degree of Master of Science from the University of Michigan. Mr. Archer then became affiliated with the research department of the Aluminum Castings Company and since 1920 he has been a member of the research bureau of the Aluminum Company of America. His work is concerned chiefly with the development and application of aluminum alloys and the metallurgical processes connected with their manufacture.

DISCUSSION

Written Discussion: By Paul D. Merica, the International Nickel Company, New York City.

If there is a school of "behaviorism" in metallurgy as well as in psychology, Mr. Archer must be a disciple of it, since the central theme of his paper is the "behavior" of a group of alloys certainly otherwise only distantly related.

He has pointed out that a certain type of "behavior" is common probably to a wide group of alloys of quite varied composition, viz, that of hardening during a tempering following a quenching operation. The cause of this hardening appears to reside in the precipitation of finely dispersed particles of a hard solid from a supersaturated solid solution of it or its constituents. Mr. Archer's general discussion of it has been so adequate that I can only suggest that he might have completed it by suggesting a suitable general name for alloy systems which behave in this manner.

There is still controversy, as Mr. Archer points out, concerning the exact mechanism of the hardening phenomenon in some cases, and we can but welcome such controversy since out of it usually emerges the truth. One stumbling block facing the "precipitation" theory of the hardening of duralumin is the established decrease of conductivity during ordinary temperature aging of it, which is difficult to imagine as the result of the change from a concentrated to a less concentrated solid solution. Mr. Archer makes some quite pertinent comments

on this difficulty and raises the question whether we have a right to expect an increase of conductivity with decreasing solid solution concentration under all circumstances.

I know of no case in which a solid solution containing very small amounts of one solute is known to have a higher conductivity than the pure metal, but cases in which solid solutions have higher conductivities than less concentrated solutions of the same solute are well established. Thus, Hunter and Sebast (*Journal*, American Institute of Metals, VII, p. 115), showed that a nickel-base alloy containing 12 per cent copper and 13 per cent chromium had a resistivity of 112 michrom-cm., whereas a more concentrated solution of 23 per cent copper and 13 per cent chromium had a much lower resistivity, 75 michrom-cm. Removal of copper from the latter has decreased the conductivity. We must not forget that a quenched piece of duralumin is probably a solid solution containing at least three components, copper, silicon and magnesium. Since minima are known to occur in the conductivity of binary, ternary and higher solid solution systems, we cannot be sure how the change in solid solution concentrations occurring during the assumed precipitation of Mg_2Si will affect the conductivity.

Furthermore, our idea of the relation between conductivity and concentration of dilute solid solution is based upon the behavior of alloy systems in equilibrium. A supersaturated solid solution is not in equilibrium; its atomic lattice is strained and it seems most probable that the electronic behavior of the atoms in such a strained lattice may be considerably different from that in a lattice which is in thermal equilibrium; the electrons may well be less tightly bound and the conductivity greater.

But there are other factors also that must play some part in the conductivity of an aged sample of duralumin. It is well known that elastic strain affects the electrical conductivity quite noticeably, tension generally diminishing conductivity in the tensional axis. A quenched bar or wire of duralumin is probably in a state of internal stress, due to thermal gradients during quenching. Upon being aged at ordinary temperature these probably are altered and quite possibly another set of internal stresses may appear due to the local straining effect of the Mg_2Si precipitation.

It appears most difficult to estimate the probable value, distribution or effect of these internal stresses but it seems most probable that they exist.

For these reasons I am compelled to join with Mr. Archer in questioning the absolute validity of the conductivity evidence against the precipitation theory. It seems that we should examine the situation at least more carefully than we have to date, before accepting it finally.

Written Discussion: By Horace C. Knerr, Philadelphia, Pa.

Mr. Archer's paper is timely in connection with the Society's recent decision to include the study of non-ferrous metals, for it contributes further evidence of the one-ness of metallurgical science.

Jeffries and Archer have made it clear that all metals obey the same fundamental laws, and that their hardness, softness, strength or ductility and the modifications which are brought about in these properties by treatment, are due, in the last analysis, to comparatively simple mechanisms. The hardening of

steel and the hardening of duralumin are but special cases of the same general group of phenomena. This knowledge greatly simplifies our problems.

It has been said that one has not mastered a subject until able to write all one knows on a half sheet of paper. Perhaps there are as yet no masters of the science of physical metallurgy in this sense, but the work of Jeffries and Archer has at least made the approach easier.

In the study of materials and forces, we first learn *what* happens,—then, after great study, we sometimes learn *why* it happens. Once acquainted with the underlying laws of cause and effect, we may apply these laws to bring about new ends, and a world of possibilities is opened. Synthetic chemistry, in this way, has produced almost miraculous results. May we not hope for a similar advance in metallurgy?

Mr. Archer has mentioned definite instances in which fundamental research and clear reasoning have, in a short time, given us several new and highly valuable metals of industry. The list could be extended.

Let us honor the pioneers in such work and lend every aid and encouragement to their efforts.

Written Discussion: By R. S. Dean, Western Electric Company, Chicago.

At the outset I wish to compliment Mr. Archer on bringing the available data on dispersion hardening of metals together in this form, so that it may be treated as a general phenomenon rather than a number of isolated examples. I think some of Mr. Archer's conclusions are open to question, however, and I will discuss several of these items in the order in which they appear in Mr. Archer's paper.

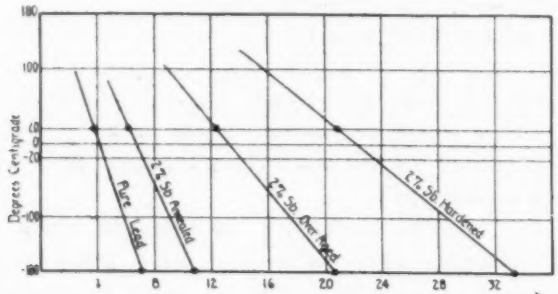
I wish first to discuss the fundamental reason for hardening by dispersion. I am not prepared to accept the so-called slip interference theory proposed by Messrs. Jeffries and Archer as a satisfactory explanation of the facts of metal hardening, although it is at once an ingenious and useful hypothesis containing, no doubt, many elements of the ultimately satisfactory theory. So far as dispersion hardening is concerned, the fundamental tenet of the theory is that particles of any intrinsically hard material will serve as a hardener if dispersed through the metal in sufficiently fine particles, regardless of any specific attraction. The hardening is, therefore, according to this theory entirely mechanical—a slip interference along the easiest paths of slip diverting the slip as Mr. Archer points out later in his paper to more complex paths.

Now there are some experimental facts which do not fit with this conception of hardening. The most cogent of these is the temperature coefficient of hardness in an alloy in the hard and soft states respectively. The accompanying figure shows the relation of Brinell hardness to temperature for a heat treated and an annealed sample of 2½ per cent antimony-lead alloy.

We see that the slope of the temperature-hardness curve is much less for the hardened alloy, so that at a temperature near the melting point the two alloys would theoretically have the same hardness. If the disperse phase exerts no effect on the nature of the forces resisting deformation, it is difficult to see how it so profoundly affects the slope of the temperature-hardness curve.

It, therefore, appears likely that there must exist a "specific attraction" between the disperse particles and the matrix to give a typical dispersion hard-

ened metal. The hardening of copper by Cu_2O and of tungsten by thoria are not entirely convincing evidence to the contrary. Certainly the amount of hardening we get with Cu_2O is not comparable to other dispersion hardenings. We have observed little hardening in the unworked state, the effect being largely



one of increased hardness due to cold work, this introduces another factor.

As an alternative to the mechanical slip interference theory, I offered some years ago (Chem. Met. Eng. *26*, 965 (1922) a theory based on the views of J. Stark regarding crystal gliding. Briefly this theory is that the ease of slip in metals is due to the fact that they are built up of positive kernels in a negative field of their free electrons. Any discontinuity of the lattice, whether introduced by gliding or a second constituent, results in a binding of these free electrons across such a discontinuity. Such a junction is stronger than the electrostatic junction of the soft metal. This is equivalent to saying that a non-polar compound has been formed across the lattice discontinuity. A development of this theory satisfactorily explains all the changes of properties observed on metallic hardening. I will not go further into this at this time, but in the near future I hope to present a comprehensive paper on this subject. There are some more specific points in Mr. Archer's paper which I wish to discuss.

The question of the relative retention of a supersaturated solution by rapid cooling and by slower cooling has been discussed in some detail by Mr. Swensen of our laboratory in connection with another paper given at this meeting. I think that there is a critical cooling velocity which retains the most solid solution. We must compromise between the greater stability and longer time given by slower cooling and the less stability and shorter length of time given by rapid cooling.

The generalization that the temperatures suitable for precipitation will be "not far below the temperatures required to cause annealing in the same alloys in the cold-worked condition" is misleading. There are several series of alloys in which the minimum precipitation temperature is several hundred degrees above the recrystallization temperature. Also we might infer that all lead base dispersion hardened alloys would "over-age" at room temperature. While this is generally the case, there are some important exceptions, so that by and large I think it is a mistake to tie up recrystallization temperature with precipitation or "over-aging" temperatures.

With regard to the checking of solubility as determined by microscopic

examination with hardening, it is certainly desirable. We have found, however, that microscopic examination is reliable if care is taken to insure equilibrium. To this end the alloys should be worked at the temperature where it is desired to reach equilibrium. In this way, equilibria have been reached in a few hours in lead alloys at room temperature, which required more than a year without the working.

The electrical conductivity of duralumin seems to be an exception. We have investigated a large number of other cases and have always found electrical conductivity the best guide to the state of the alloy. It should be possible theoretically for a disperse aggregation to have a greater resistance than the corresponding solid solution, even if the dispersion were uniform, but such an alloy would certainly be practically glass hard.

Written Discussion: By O. W. Ellis, Westinghouse Electric and Manufacturing Company, East Pittsburgh, Pa.

The author has rendered a signal service to metallurgists in collating the results of the considerable amount of work that has been done on the subject of the hardening of metals by dispersed constituents precipitated from solid solutions.

In this connection the author has had occasion to refer to the work carried out by the writer and D. A. Schemnitz on the hardening of the beta solid solution of the copper-zinc system by dispersed alpha solid solution. The author has hinted, though he does not directly state, that the writer and his associate did not accept the "key" hypothesis as an explanation of the observed increase in the hardness of the samples tested by them. The author's opinion is apparently based on the somewhat ambiguous statement in the writer's paper —"It is difficult to imagine how soft particles can act as "keys" to prevent the relative slip of the atom bearing planes of a solid solution." The difficulty in the mind of the writer and his associate, however, lay not in the acceptance of the "key" hypothesis. They were firm believers in this concept. For his associate the writer cannot speak, but for himself he can say that he still accepts the "key" hypothesis as best explaining the observed facts. What was found difficult by the writer and Mr. Schemnitz was to explain the 12.8 per cent increase in the hardness of beta solid solution occasioned by the precipitation of *soft* keys of alpha solid solution. It is not impossible that the 12.8 per cent increase in hardness can be explained by assuming "that in order for a precipitated particle to act as an effective key it is only necessary that its resistance to deformation in a random direction be greater than the resistance to deformation along a slip plane of the matrix in which it is imbedded." But the writer and his associate felt that the observed increase in hardness could not wholly be accounted for by such an assumption. This was entirely a matter of opinion. They had no firm basis on which to support that opinion. Having this opinion, however, it was natural that they should seek for an explanation of the observed increase in hardness. Part of this they believed to be due to such a cause as the author has suggested. But the larger portion of the increase in hardness they felt to be due to the precipitation of the solute (alpha solid solution) in the form of submicroscopic spherical particles (of the nature of colloidal particles) whose boundaries were controlled by surface

tension and which, on this account, were specially resistant to shearing forces and consequent internal slip along gliding planes. Further, they assumed that these spherical particles had a distorting effect upon the space lattice of the beta solid solution in their immediate neighborhood. On aging the alloy, these spherical particles, it was assumed, changed into minute allotriomorphic crystals when the forces of crystallization of the alpha solid solution exceeded those surface-tension forces that controlled the boundaries of the spherical particles.

On the above basis the writer and Mr. Schemnitz explained the phenomenon of the hardening of beta brass, an explanation which can be applied to other cases of the hardening of metals by dispersed constituents precipitated from solid solution, as follows: "While spherical particles, whether soft or hard, are in process of formation the hardness of the alloy will increase, because of the increasing resistance offered to slip . . . as a result of the precipitation of particles of this form. When the spherical particles begin to change into minute allotriomorphic crystals the hardness of the alloy will tend to decrease and will actually be lowered (1) when the rate of formation of allotriomorphic crystals exceeds the rate of formation of spherical particles and (2) as the allotriomorphic crystals themselves increase in size." (Quotation from paper by Ellis and Schemnitz: American Institute Mining & Metallurgical Engineers, No. 1348 N, 1924.

While speaking of the hardening of quenched alpha-beta brass on "aging" it may be of interest to refer to later work by the writer, which has been withheld from publication pending an opportunity for fuller investigation of certain of the phenomena that were observed. The writer's later work has consisted in the measurement of the electrical resistivity of quenched alpha-beta brass on slow heating up to 500 degree Cent. What has been noted is that the electrical resistivity of quenched alpha-beta brass, i. e., brass containing from about 39.7 to about 46.7 of zinc (not 37 to 44 per cent of zinc, as suggested by Fig. 4 of the author's paper), increases uniformily up to temperatures varying from

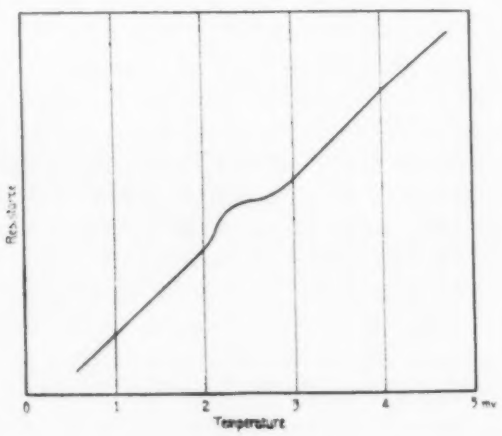


Fig. A

265 to 300 degrees Cent., at which temperatures the resistivity is subject to a marked advance. Reference to Fig. A, which is a typical resistivity-

temperature curve of the type now under discussion (actually that for the 60.8 per cent copper alloy), will make the above statement quite clear. The change in the coefficient of resistivity occurs at temperatures which appear to be higher the greater the content of zinc in the brass or, to express the same thought in other terms, at temperatures which appear to increase inversely as the proportion of alpha that can be precipitated from the beta solid solution. It appears reasonable to assume that the increase in resistivity is occasioned by the precipitation of alpha solid solution in a dispersed state (cf. the work of Homerberg and Shaw—*Transactions*, Institute Mining and Metallurgical Engineers, 1924, 70, 365). The temperature at which the change in resistivity is complete varies from 360 to 400 degrees Cent., i. e., the range of temperature over which abnormal resistivity values, if we may so term these values, are obtained occupies about 100 degrees Cent. The inversions were most marked in

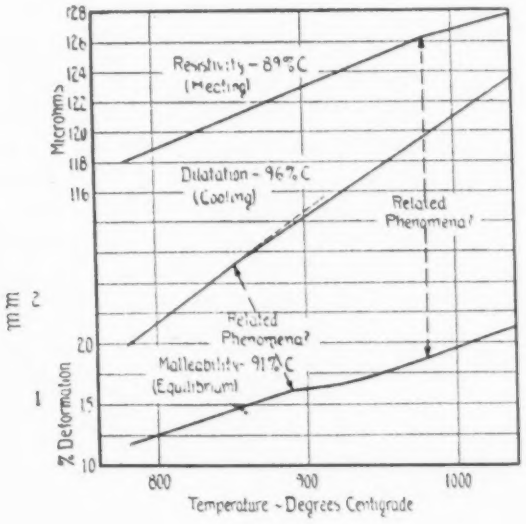


Fig. B

the alloys of lower zinc content, i. e., in those alloys from which the greatest amount of alpha could be precipitated. The phenomenon was not observed in an alloy containing 61.8 per cent of copper, but was noted in alloys containing from 60.8 to 55.8 per cent of copper. It is probable that the reason for the non-appearance of the inversion in the resistivity-temperature curve for the 61.8 per cent copper alloy lay in the fact that it was found impossible to prevent precipitation of alpha by quenching.

In concluding his contribution to this discussion the writer desires to call attention to a phenomenon which has received but little consideration, but which appears well worthy of note. The writer has in mind the well-authenticated change in the electrical resistivity of steel which occurs at about 980 degrees Cent., i. e., when the steel is in austenitic condition. The character of this change may be gathered by reference to Fig. B, in which the highest curve represents the resistivity-temperature relationship for a 0.89 per cent carbon steel (after Sodau-Carnegie Scholarship Memoirs, Iron Steel Institute, 1916, 7, 195). This change in electrical resistivity appears to be connected with

changes that have been observed in the volume and the malleability of steel in the austenitic state. In the formr case, that of volume, Andrew, Rippon, Miller and Wragg *(Journal,* Iron Steel Institute, 1920, 101, 527) have shown that steel is subject to a slight diminution in its ceofficient of dilatation at temperatures varying from about 920 to about 860 degrees Cent., the temperature of this change point being lower in the case of steels of higher carbon content. The character of this change will be appreciated on reference to the middle curve of Fig. B, which represents the volume-temperature relationship for a 0.96 per cent carbon steel; (after Andrew and his collaborators). In the latter case, that of malleability, the writer has found that steel is subject to a decrease in malleability at temperatures which agree very closely with those at which Andrew and his co-workers have noted the decrease in volume. The character of this change will be understood on reference to the lowest curve of Fig. B, which represents the malleability-temperature relationship for 0.91 per cent carbon steel.

A comparison of the three curves of Fig. B, which might be applied with slight changes to practically all the hypoeutectoid steels, will make it clear why the writer is of the opinion that an alteration in the character of the solid solution of carbon (no matter in what form it occurs) in gamma iron ensues at temperatures well within the austenitic area of the iron-cementite diagram. He is of the opinion that at the temperature at which the changes in volume and malleability occur there is an alteration in the degree of dispersion of the solute in the austenite. This alteration in the dispersion of the solute results in an increase of density and of hardness. The change in hardness is not fully complete until a temperature of about 1000 degrees Cent. is reached, at about which temperature the austenite suffers a change in electrical resistivity. The decrease in resistivity actually coincides with a decrease in the hardness, (an increase in the malleability). This change is doubtless connected in some way with the observed increase in the grain size of austenite which occurs at about 1000 degrees Cent. This increase in the grain size of the austenite, the writer believes, is occasioned—permitted, one might almost say—by the alteration in the dispersion of the solute in the austenite.

Here is an almost untouched field that awaits investigation, a field closely bordering that which has been the subject of the author's interesting monograph. The application of X-rays to the study of these interesting phenomena, in particular, appears to be well worthy of consideration. Whether the study of this field will prove fruitful remains to be seen, but that such study would be worth while the writer feels to be quite clearly evident.

Written Discussion: Dr. Marie Gayler, National Physical Laboratory, Teddington, England.

Dr. Marie Gayler has read with interest Mr. Archer's paper on the theories of the hardening of metals and analysis of their heat treatment for hardening. On page 726 the author refers to unpublished results obtained by P. V. Faragher verifying the fact "the prolonged heating at a low temperature gives greater hardness than short heating at a higher temperature", for aluminum-copper alloys both with and without magnesium. The writer is carrying on a systematic research on age hardening of aluminum alloys of various compositions

and has also found similar results. The temperature and time of heat treatment at that temperature bear a definite relation to the hardness obtained; and the higher the aging temperature, the shorter the period of heat treatment. The maximum hardness obtained by such heat treatment varies according to the temperature and it does not necessarily follow that the higher aging temperature gives the maximum hardness. The writer's results, as yet unpublished, indicate that the temperature of heat treatment which will give the maximum hardness possible in a short time is fairly critical and depends on the previous condition of the alloy.

On page 729 the author notes that marked hardening was observed in a magnesium alloy containing 6 per cent aluminum and, assuming the validity of the precipitation theory of hardening, suggests that the solubility of aluminum in solid magnesium at room temperature is less than 6 per cent. The present writer agrees with the author regarding this assumption. When Dr. Hanson and she published their Aluminum-Magnesium equilibrium diagram, the method they used for the determination of the limits of solubility, was that of slow cooling the alloys previously annealed for several hours at a high temperature, down to the required temperature and quenching. The alloys containing 2-12 per cent aluminum after 62½ hours at 430 degrees Cent. were slowly cooled to 150 degrees Cent., over a period of 7 hours only. The writer and Dr. Hanson have since found that the rate of cooling depends on the constitution of the alloys in question, for in some cases they found that slow cooling over a period of weeks was necessary for true equilibrium conditions. It is therefore most probable that the rate of slow cooling used for the determination of the solubility limits of aluminum in magnesium was too fast and that a slower rate of cooling would appreciably shift the solubility line over towards the magnesium end of the equilibrium diagram.

The present writer was much interested in the new information given by the author on page 740, relating to the fact that when extremely pure aluminum is used, copper-aluminum alloys containing about 0.5 per cent magnesium and 0.02 per cent silicon, give as great an age hardening effect after suitable heat treatment as ordinary duralumin. In a paper on the "Ternary Alloys of Aluminium with Magnesium and Copper"*, the present writer attributed the age hardening which took place in alloys containing more than 1 per cent copper and less than 2 per cent magnesium, to the unavoidable presence of Mg_2Si and from the constitution of these ternary alloys, she assumed that no age hardening would take place in the absence of silicon in the aluminum. It is extremely improbable, as the author states, that Mg_2Al_3, $MgCu_2$ or $CuMg_4Al_6$ are precipitated during hardening since from the constitutional diagram of the ternary alloys the $(Al + CuAl_2)$ phase field exists up to 8 per cent magnesium. The author's remarks are therefore extremely interesting and the results obtained, though inexplicable at present, may throw further light on the phenomenon of age hardening.

Author's Reply

The author wishes to express his appreciation of the interest with which

* Gayler, M. L. V., *Journal*, Institute of Metals. Vol. 29, No. 1, 1923.

the paper has been received and of the unusual value of the written discussions.

He would like to comply with Dr. Merica's suggestion by proposing a "suitable general name" for alloy systems of the type described, but a name which is at the same time sufficiently descriptive and brief is quite difficult to find or coin. Furthermore, the author felt that he was perhaps offering enough for the present in the way of new terms in "solution heat treatment", "precipitation heat treatment" and "precipitation hardening". The term "age hardening" has already gained wide usage and begins to trip easily from the tongue. Alloy systems exhibiting this phenomenon may well be referred to as "age hardening systems". This term can probably be applied properly only to systems which harden on natural or ordinary temperature aging. A more general term would be "precipitation hardening systems". After some practice, this term might seem a little less awkward than at first.

Merica's suggestions regarding the electrical resistivity difficulty are interesting, and the author is glad to note Merica's agreement that this difficulty can scarcely be considered as serious evidence against the validity of the general theory of precipitation hardening. In regard to the nickel base alloys described by Hunter and Sebast, the present author would consider it possible that the addition of 23 per cent copper might throw some chromium out of solid solution, thus accounting for the decreased electrical resistivity. Formerly, nickel and chromium were commonly considered to be mutually soluble in all proportions. It was only by the application of X-Ray crystallography that the apparently homogeneous grains of some of these alloys were found to contain two different space-lattices. Unless this method has been applied to the nickel-copper-chromium alloy referred to, there would seem to be some question that it is a homogeneous solid solution.

The author must confess failure to see the cogency of the temperature-hardness relations in Mr. Dean's lead-antimony alloy as evidence against the key hypothesis. At a temperature near the melting point, the originally annealed and hardened specimens would have the same constitution and structure—a homogeneous solid solution—except perhaps for some difference in grain size. They would therefore have substantially the same hardness. Since their initial hardnesses are quite different, it seems obvious that their temperature coefficients of hardness *must* differ.

The hardening effect of Cu_2O on metallic copper is of course relatively slight since the oxide particles are ordinarily much larger than those required to produce large hardening effects.

Mr. Dean's promised paper on the theory of hardening will be awaited with great interest. From the brief outline given, it would seem that the hardening caused by dispersed particles should be independent of the specific hardness of such particles, since "any discontinuity of the lattice—results in a bonding of these free electrons" and a stronger junction. This is of course a vital point, since one of the chief values of a theory of this phenomenon should lie in pointing the way to the kind of particles from which hardening may be expected.

The author notes Dean's disagreement with the correlation of precipitation and recrystallization temperatures. It is to be hoped that Dean will present

detailed information concerning the alloys in which the precipitation temperatures are "several hundred degrees above the recrystallization temperatures." The generalization proposed by the author is of course tentative and admittedly only approximate at best. There does seem to be some physical basis, however, for correlating the two classes of phenomena, since both represent the fairly sudden change from an unstable condition to one of greater stability. It seems that the temperatures required to overcome the inertia of the predominating metal or solid solution should be about the same whether the change is precipitation or recrystallization. It has been pointed out previously that higher temperatures are required for the diffusion of solute elements of high atomic volume, than for elements of low atomic volume.

The author notes with interest that Ellis is in general agreement with the key hypothesis. Ellis' idea that the smaller precipitated particles are more resistant to shear than the larger ones, due presumably to less perfect crystallinity at the surface, is an interesting suggestion.

The solubility values given by Ellis for the copper-zinc system are undoubtedly more accurate than the older values from which the author's diagram was taken.

The little understood changes in austenite to which Ellis has referred would seem to warrant investigation, as he suggests. It is possible that they are caused by changes in the non-metallic constituents rather than in the distribution of carbon. Attention should be given, at least, to the possibility of such other causes.

The author is pleased to note Dr. Gayler's interest in the paper, her confirmation of some of the aging characteristics of aluminum alloys, and her agreement that the solubility of aluminum in magnesium at room temperature is probably less than indicated in the diagram by Dr. Hanson and herself.

DENDRITIC CRYSTALLIZATION AND GRAIN FORMATION IN STEELS

By Dr. V. N. Krivobok

Abstract

This paper deals primarily with the "secondary" transformations which take place in steels in the solid state and which, in the author's opinion are extremely important. The author sets forth his observations made in this field and feels that his findings may be of assistance to the man actually engaged in steel making.

The generally accepted ideas of the mechanism of crystallization are briefly discussed.

That dendritic segregation is not the sole cause of defective steel and a pronounced dendritic structure is not sufficient cause for the rejection of steel, is one of the conclusions arrived at by the author from his observations. He is of the opinion that the process of solidification has not received the proper amount of attention.

By means of photomicrographs the writer shows the microstructure of dendritic branches found in steels of different composition and the non-metallic inclusions ordinarily found in steels and those found in steels of large and well-developed dendritic structures. The results of various heat treatments, quenching, heating and cooling upon the microstructure of steel are also shown.

THE process of solidification and crystallization of metals has been extensively investigated and in its simplest form is fairly well understood. In many cases, however, the primary process of crystallization is followed by complicated transformations which take place immediately after solidification. In steels these "secondary" transformations, taking place in the solid state, are extremely important. The purpose of this paper is to present the results of some observations in this field.

It is hoped that the information about to be presented will serve a useful purpose, namely, to be of assistance to the man actually engaged in steel making.

Before presenting the observations on secondary transforma-

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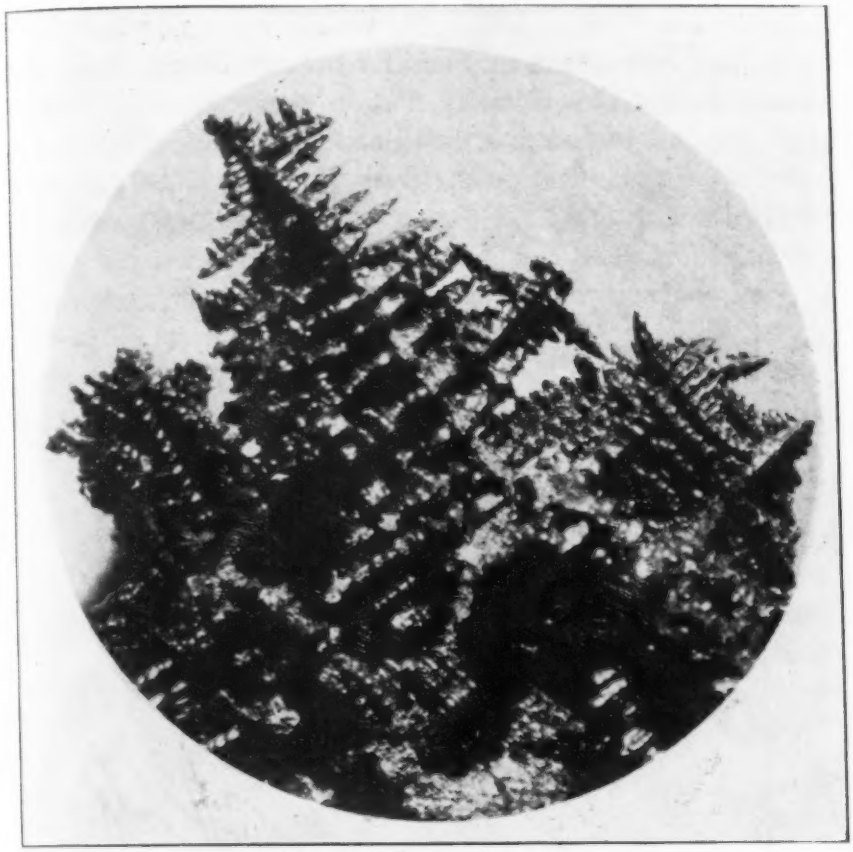

Fig. 1—A Dendritic Skeleton. 10x.

tion, it is considered advisable to discuss briefly the generally accepted ideas of the mechanism of crystallization.

Upon reaching the solidification temperature, a metal or an alloy begins its transition from the liquid to the solid state, through the formation of a solid crystalline particle somewhere within the melt, at a point known as a "nucleus." Although the nucleus may have no physical actuality, it is the term by which we denote the place at which, for reasons not altogether clear to us, the first solid particles choose to appear. As a rule, the process of solidification proceeds quite rapidly, unless special precautions are taken to retard it. Development about the nucleus proceeds at first in one direction only, resulting in a particle being developed in one direction, instead of growing in all directions. We say that the formation of the "first axis" has taken place. At the next instant, other axes will be developed, which, as a rule, are perpendicular to the first. If we could withdraw at this moment all the

metal that has not yet solidified, we would find a number of beautiful crystalline formations very similar to snow flakes. It is not easy to remove the metal artificially, but in some so-called "shrinkage cavities" we find just such crystalline formations. In Fig. 1 is reproduced, at a magnification of 10 diameters, a picture of a rather rare crystalline formation. For reasons to be given later, such a

Fig. 2—A Dendritic Skeleton. 3x.

formation is called a "dendritic skeleton." It was found in the shrinkage cavity of an ingot of commercial steel of the following composition:

	per cent
Carbon	0.62
Manganese	0.37
Silicon	0.28
Phosphorus	0.003
Sulphur	0.02

A similar dendritic skeleton was found in another ingot of practically carbonless iron (carbon 0.05 per cent). This skeleton was mounted in a low melting point alloy and then polished so as to offer a suitable surface for microscopic examination. A section through this dendritic skeleton is reproduced in Fig. 2, at a mag-

nification of three diameters. Dendritic skeletons are found in special steels as well as in plain carbon steels.

In the later stages of the process of solidification, the further development of the dendritic skeletons into more complete crystalline units takes place. The new additional metal is deposited on the axis or branches rendering them thicker, while it is probable that new branches may also be formed. Such a procedure will go on until the spaces between the branches are completely filled. At this stage the *dendrite* is completed and the solidified part of the ingot will be made up of a large number of individual dendrites, necessarily idiomorphic.

The dendrites have certain characteristics, namely:

1. They are fully developed crystals, idiomorphic in shape and heterogeneous in chemical composition (except in pure metals.)
2. They are of a size quite visible to the unaided eye; the size is dependent upon the method of cooling.
3. They are crystallographic units at the time of their formation.

When a section of steel sample is properly polished and etched by a suitable reagent, the dendritic formation will usually be revealed. Such a formation is known as "dendritic structure." The appearance of this structure in alloys is due to the mechanism of solidification, during which the "progressive" solidification with varying composition takes place, thus producing dendritic segregation. Dendritic structure is, therefore, merely the visible manifestation of dendritic segregation.

Dendritic formation (during the solidification of an alloy) occurs *always*. It cannot be suppressed. The dendritic structure can be prevented if we succeed in preventing, by artificial means, dendritic segregation.

If we could, by some selective method, treat a sample of a steel in such a way as to bring out only the boundaries between the individual dendrites, we would obtain a structure of very large grains, in fact so large as to be visible without the use of a microscope.

When we examine a sample of steel, etched in the usual way (with a dilute solution of nitric acid in alcohol), we find it is composed of a number of small units which are generaly called

"grains." We speak of steels as having "coarse" or "fine" grain structures. These grains, even if they belong to the class of "coarse" grains, are always much smaller than the areas occupied by the finest dendrites. The exceptional case of artificial grain growth is, of course, excluded.

As was mentioned before, each dendrite is, at the time of its formation, a physical as well as a crystallographic unit. That is to say, it has the same crystallographic characteristics throughout. At room temperature, each grain has its own definite crystallographic arrangement, no matter how fine the grain may be. Bearing this in mind, and remembering what has been said about the size of grains, we must consider as quite probable the modern theory which states that upon cooling from below the solidus line down to room temperature, the body of each dendrite divides into a number of parts, to which the name of "grains" has been given. Colonel Beliaew calls this process of subdivision the process of "granulation."

In support of the theory of granulation, which, in spite of much evidence, is not universally accepted, this paper presents some additional confirmation.

Structure of Dendritic Skeletons

As mentioned before, the sectioning of a number of dendrites would present, upon proper preparation, a simple structure; it would consist of a number of areas with irregular outlines, each area representing a section through a dendrite. These areas would be much larger than the ordinary grain size, but this can be hardly considered as sufficient evidence for the theory of granulation.

It is obvious that we have no method of distinguishing between grains which have resulted from the formation of dendrites, and those that were developed during the process of granulation.

The only way in which such a differentiation could be made would be to secure an individual dendrite or a dendritic skeleton, and examine its structure. The question of the existence of granulation could then be settled.

In Fig. 2 is shown a section through a dendritic skeleton found in practically carbonless iron. This section offers the opportunity of examining the structure of the metal of which dendritic branches are made. This metal represents a part of the

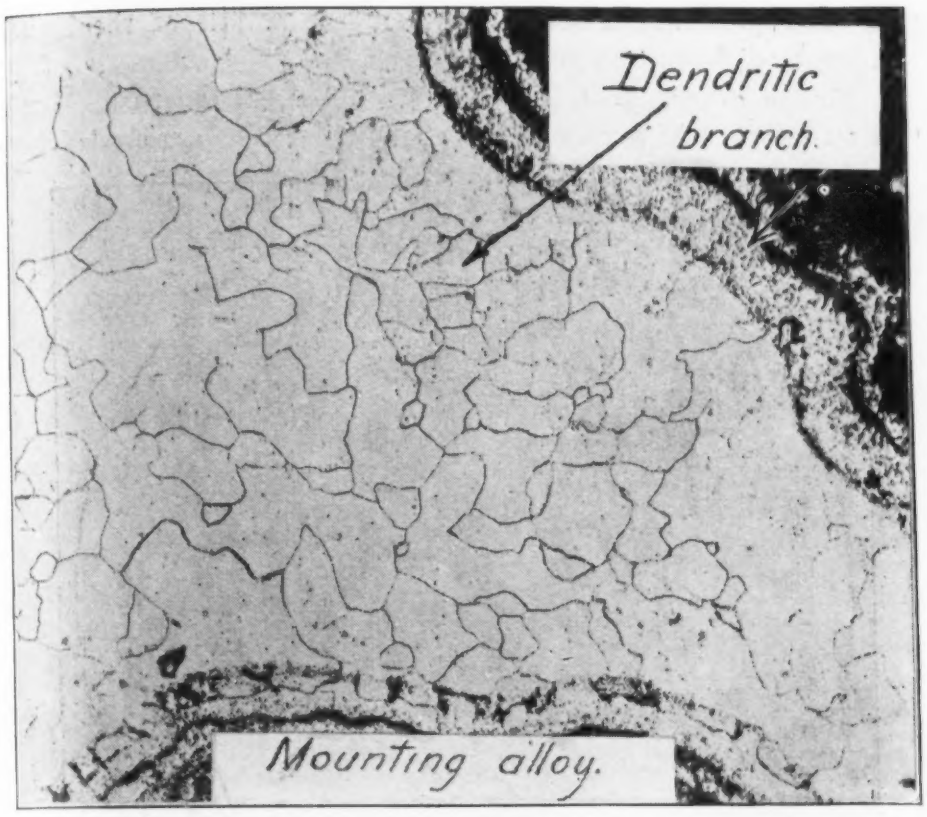

Fig. 3—A Section Through a Dendritic Branch; Etched with Nitric Acid. 200x.

individual dendrite which, if undisturbed, would have grown to completeness, and would have formed a part of the steel ingot. In other words, it gives us the opportunity to examine the structure of an infinitely small body of steel made up of a part of a dendrite.

The sample shown in Fig. 2 was etched with 5 per cent nitric acid in the usual way. It was found to possess a structure absolutely similar to the structure of an ordinary commercial ingot of similar composition. (See Fig. 3). It consists of a number of polyhedral grains, differing in size and shape. This section has been taken from the spot a-a and shows the complete section through the dendritic branch. The structure of the whole ingot in the cavity of which the dendritic skeleton was found is shown in Fig. 4. It is in no way different from the structure of the dendritic branch.

In Fig. 1 is shown a picture of quite a remarkable dendritic skeleton. A part of the skeleton was detached and mounted so

that after polishing, a section through the middle of the branch was exposed for examination.

Photomicrograph of Fig. 5 shows the microstructure. Again this structure, which consists of pearlitic grains surrounded here

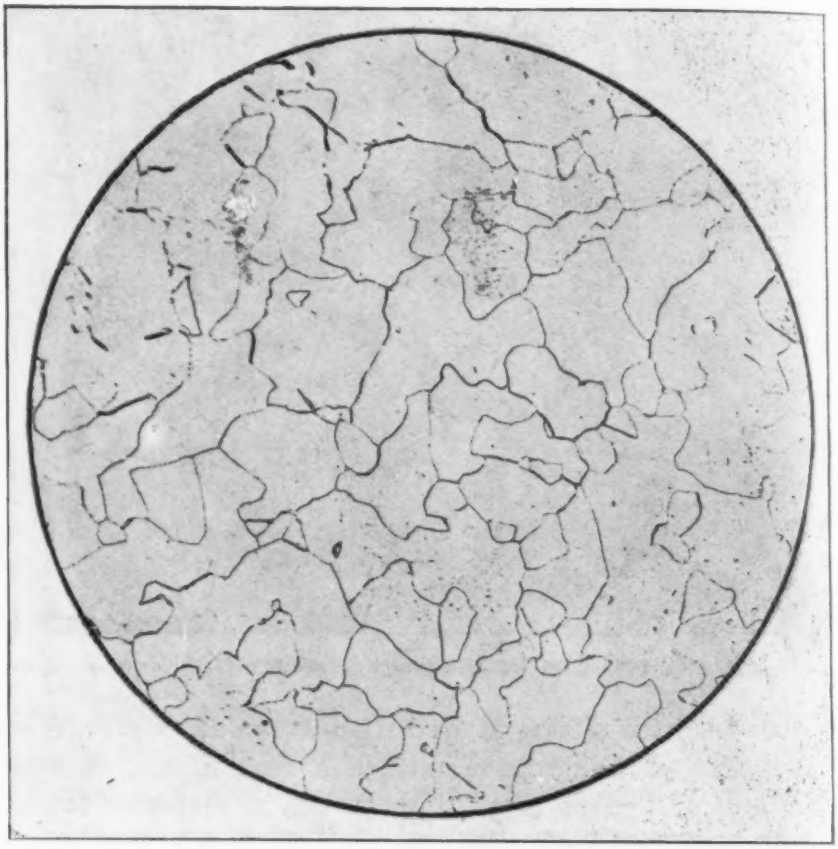

Fig. 4—The Microstructure of the Ingot in the Cavity of which the Skeleton Reproduced in Fig. 2 was Found. Etched with Nitric Acid. Mag. 200x.

and there by ferritic network, is in no way different from the structure of any other part of the ingot which had the same composition.

A dendritic skeleton of eutectoid composition was next studied. In the photomicrograph of Fig. 6 is shown a number of dendritic branches with their microstructures. The nature of the structure (namely that of eutectoid composition) can be clearly seen. Photomicrograph, Fig. 7 is of particular interest; here four branches can be seen separated from each other by a very thin space containing no metal. These branches were apparently just about ready to close up and form a piece of perfectly solid steel.

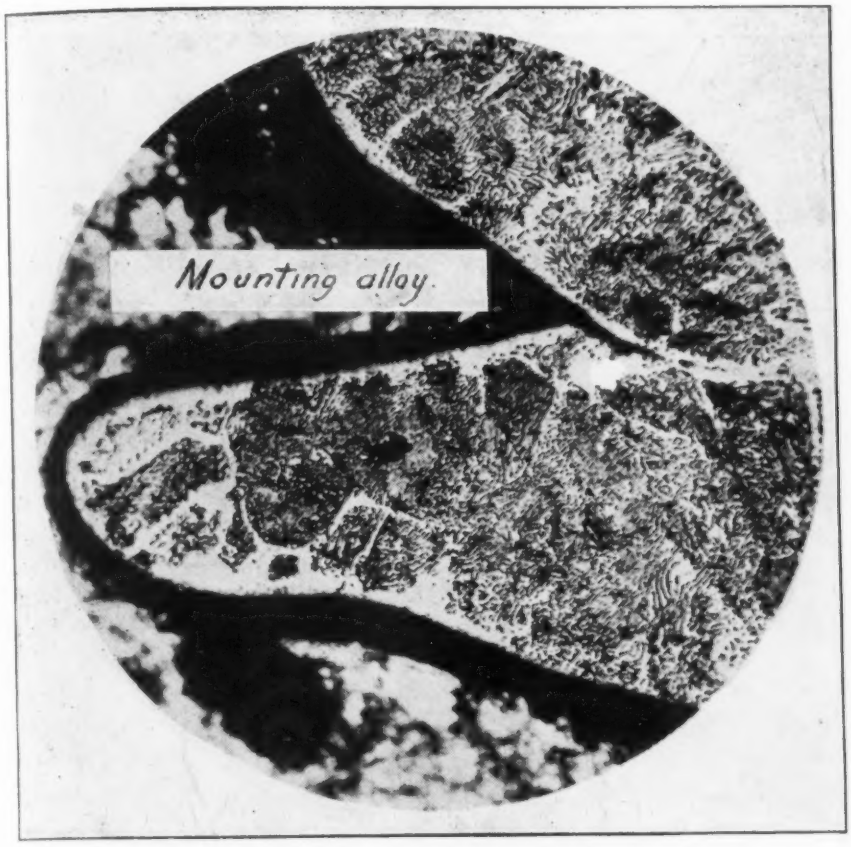

Fig. 5—The Microstructure of the Dendritic Branches Detached from the Skeleton which was Shown in Fig. 1. Etched with Nitric Acid. 200x.

The structure of still another sample is of interest. The composition is as follows:

	per cent
Carbon	0.27
Manganese	0.65
Silicon	0.19
Sulphur	0.023

It was taken from the shrinkage cavity of an ingot which had a few dendritic skeletons adhering to the surface of the metal. A sketch of this skeleton is shown in Fig. 8 and the place at which the photograph was taken is indicated. The photomicrograph, Fig. 9 shows the typical structure of steel in "as cast" condition.

From the examination of isolated dendritic skeletons, we have found that the branches of the dendritic skeletons, which were *crystallographic units* when formed, show upon cooling to room temperature, a multi-grained structure, each grain possessing its own characteristic orientation. Consequently, a branch of den-

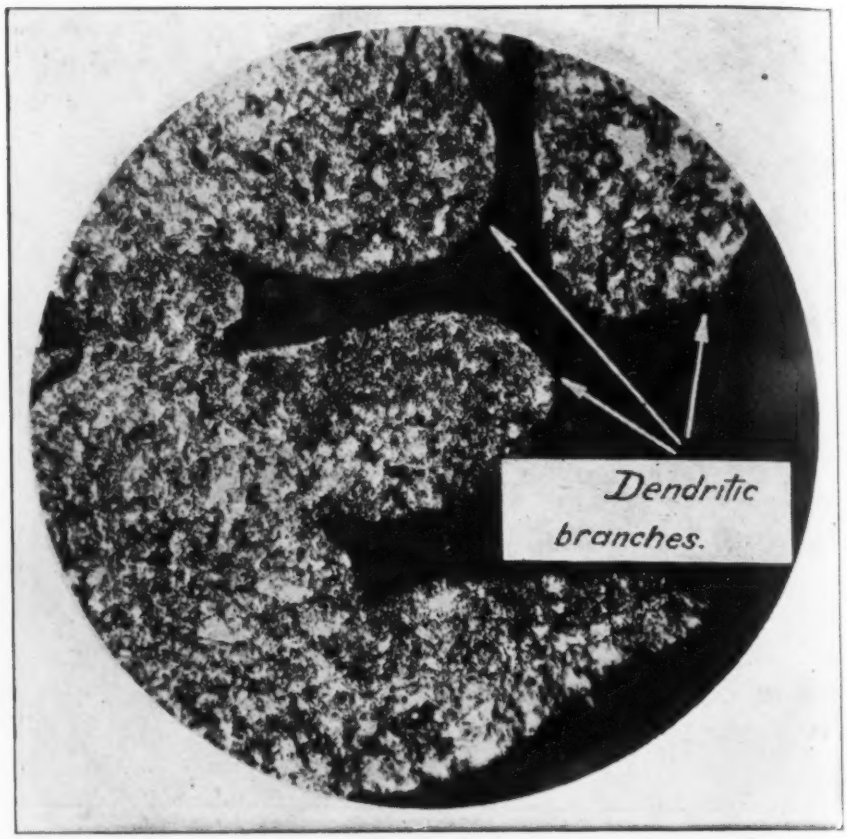

Fig. 6—The Microstructure of the Dendritic Branches Found in Steel of Eutectoid Composition. 200x. Shows Four Branches. Etched with Nitric Acid.

dritic skeleton is no longer a crystallographic unit although it still preserves its original outer form. It is now composed of *a number* of small crystallographic units.

The study of the inner structure of dendritic skeletons furnishes, it seems, sufficient evidence for the theory of granulation. This study indicates that through a certain process a single dendritic skeleton or a single dendrite (crystal) is changed (subdivided) into a number of grains, which are those observed under the microscope. It is obvious that if such a process had not taken place, the inner structure of the dendritic skeletons could not be observed.

While any name may be used to denote the process of subdivision, "granulation" is the name proposed by Colonel Beliaew and is perhaps as appropriate as any other.

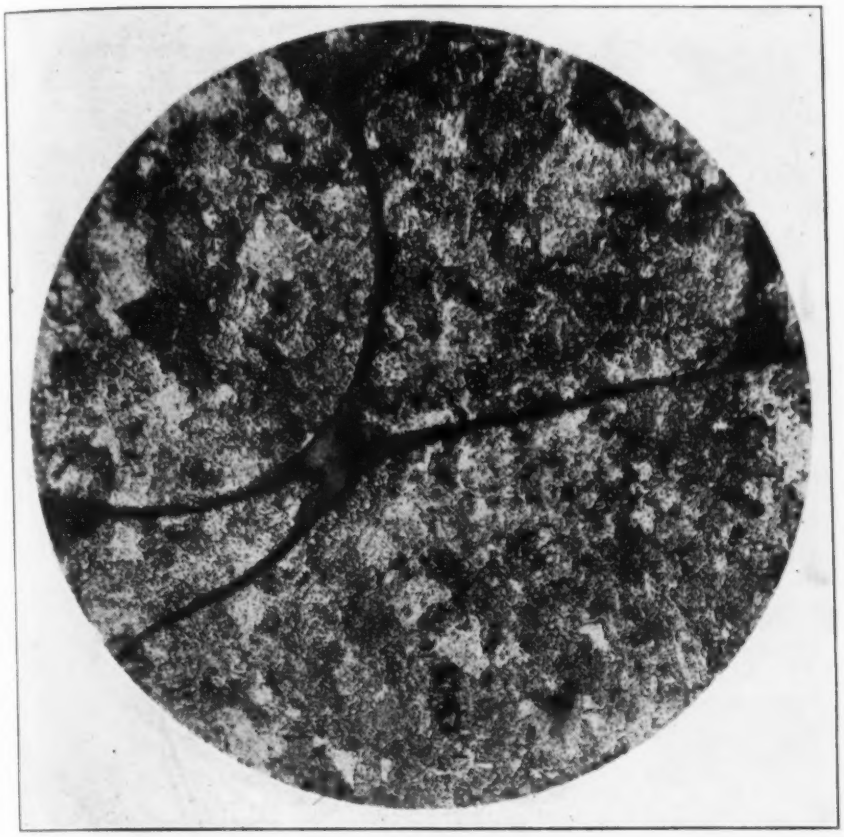

Fig. 7—Shows Four Branches of a Dendrite Just About Ready to Close Up and Form a Homogeneous Piece of Steel. 200x. Etched with Nitric Acid.

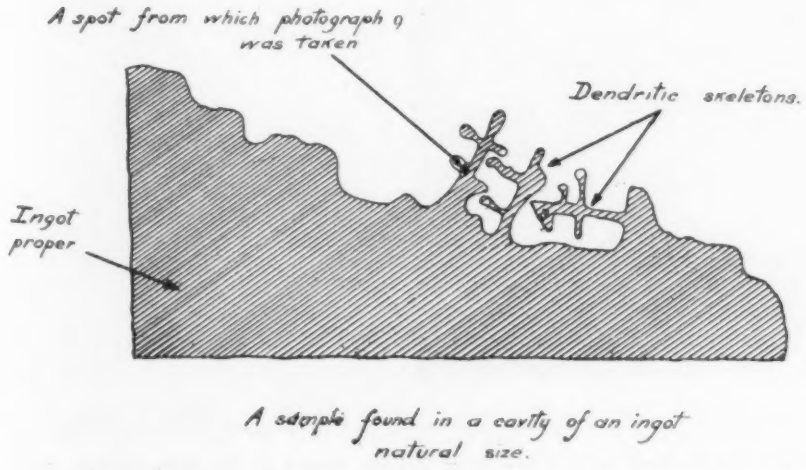
A spot from which photograph 9 was taken
Dendritic skeletons.
Ingot proper
A sample found in a cavity of an ingot natural size.

Fig. 8—Sketch Showing the Cross Section of a Dendritic Skeleton Found in the Cavity of an Ingot.

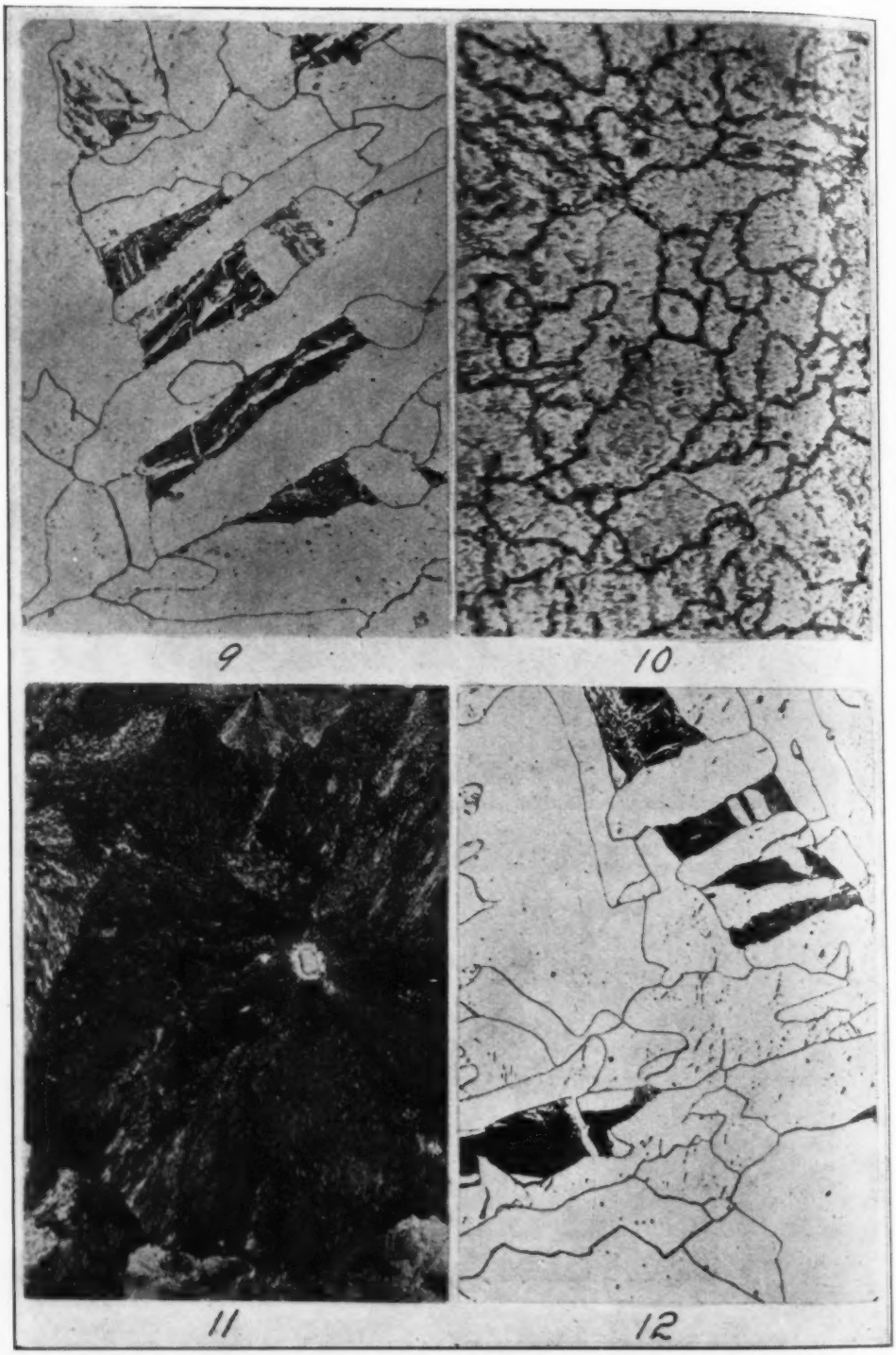

Fig. 9—The Structure of a Dendritic Branch Found in the Low Carbon Steel. 200x. Etched with Nitric Acid. Fig. 10—A Structure of 0.50 Per Cent Carbon Steel Heated in Vacuo to 75 Degrees Above Critical Point and "Air" Etched. 200x. Fig. 11—This Photograph Illustrates an Observation that in Many Instances the Grains of Steel Grow Around an Inclusion. 350x. Etched with Nitric Acid. Fig. 12—Structure of Steel in which Dendrites Illustrated in Fig. 9 were Found. 200x.

Fig. 13—Microstructure of Steel which Possesses a Very Large Dendritic Structure. 200x. Etched with Nitric Acid. Fig. 14—Microstructure of Steel with Dendritic Structure of Small Size. 200x. Etched with Nitric Acid. Fig. 15—Non Metallic Inclusions as Ordinarily Found in Steels. 200x. Unetched. Fig. 16—Non Metallic Inclusions as Found in Steels of Large and Well Developed Dendritic Structure. 200x. Unetched.

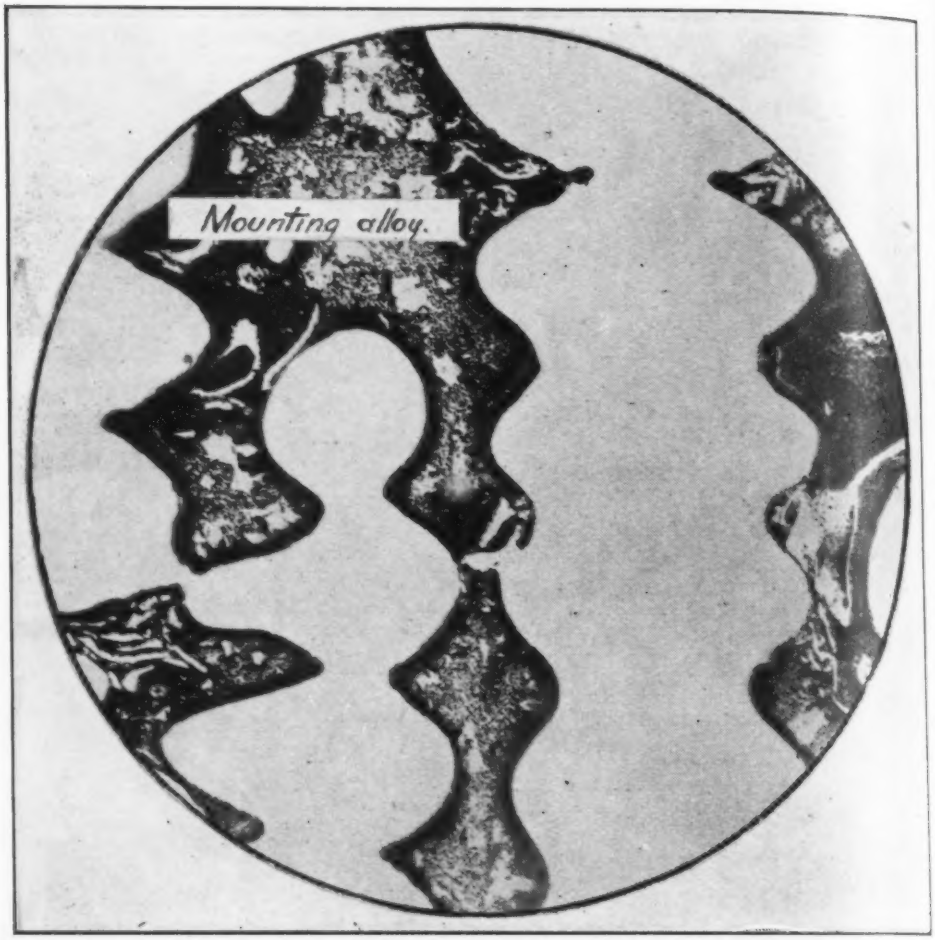

Fig. 17—A Section Through Dendritic Branches Mounted in the Low Melting Point Alloy. Unetched. 100x.

Causes of Granulation

It is not easy to find an answer to the question, "Why does granulation take place?" This question, so it seems, is closely related to another, namely, "Where (on the equilibrium diagram) does granulation occur?" and perhaps it will be appropriate to consider this last question first.

It is generally accepted that granulation occurs on the part of the diagram between the "solidus" line and the upper critical point; that this transformation is not sudden but gradual; and that the number and the size of the grains can be artificially controlled.

The writer has endeavored for a long time to secure more evidence on this point. The change from one allotropic form to

Fig. 18—Segregation of Non-Metallic Impurities in Otherwise Clean Material. 100x. Unetched.

another, might be considered as one of the causes that bring about granulation. If such were the case then granulation should occur rather suddenly and at a temperature near that of the upper critical transformation point. However, as we are well aware, mere heating to the temperature of the critical transformation never results in the change of structure.

A piece of steel of approximately 0.50 carbon was heated to temperatures 25, 50, and 75 degrees above its Ac_3 transformation and was allowed to cool as slowly as possible. Heating and cooling were carried out in vacuo. As soon as the sample had reached the desired temperature, a small amount of air was allowed to enter the furnace and produced an effect much like etching by acids at elevated temperatures.

As long as the temperature was 25 or 50 degrees above the

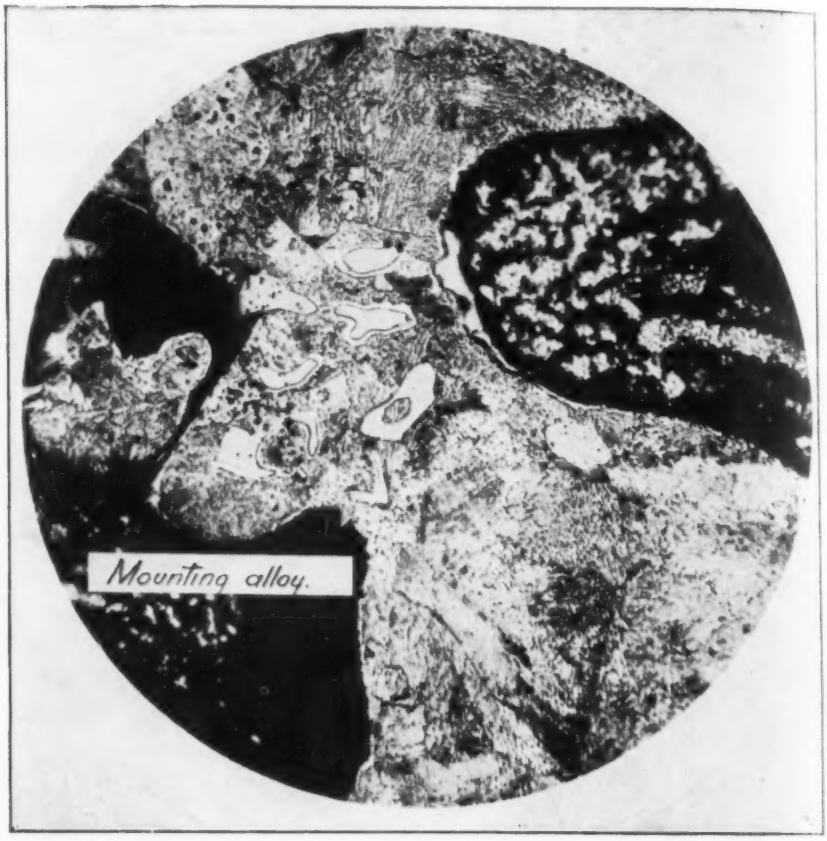

Fig. 19—Same as Fig. 18. Very Bad Segregation. Etched with Nitric Acid. 100x.

upper critical point the structure of the sample as revealed by this "air etching" was unchanged. At a temperature 75 degrees above the critical range, a noticeable change occurred. The sample showed a marked polygonal structure, much like that of austenite, (photomicrograph, Fig. 10) as revealed by etching at high temperature.

It seems probable that a change in structure accompanies the change in atomic arrangement (Ac_3 transformation), but that the structural change is much slower and consequently a certain time must elapse before the effect of the atomic rearrangement is observable by the microscope. A temperature range of 75 degrees is not inconsistent with this explanation.

There are other possible causes for granulation besides allotropic change. First may be considered the presence of internal stresses. The author's recently published work[1] on the recrystal-

[1]American Institute of Mining and Metallurgical Engineers., February, 1926.

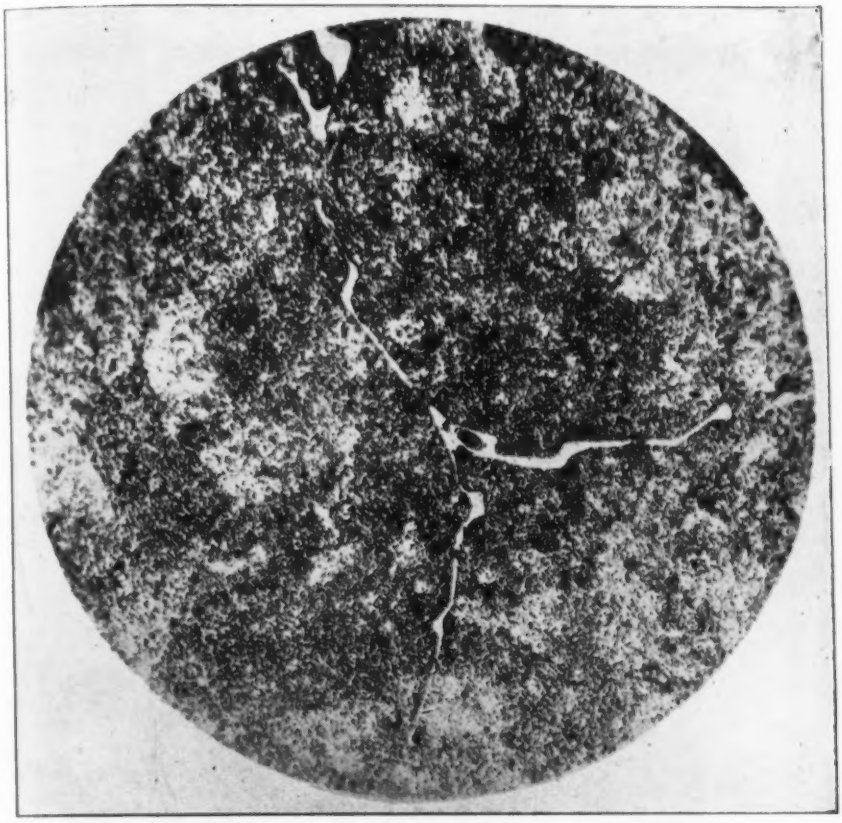

Fig. 20—Non-Metallic Impurities Forming a Continuous Film in Steel. This Steel Contains Very Large Dendrites. Its Microstructure has been Refined by a Heat Treatment. 200x. Etched with Nitric Acid.

lization of single crystals of iron-silicon alloy indicates that a single crystal (which is in reality a complete dendrite isolated from the rest of the dendrites) can easily be converted into an agglomeration of much smaller grains, provided the crystal has been strained and heated to a certain temperature. Granulation could not be produced and the structure remained unchanged in a single crystal of this alloy, when it was detached from the ingot and heated by itself without straining. However, if two or more of these single crystals were heated without separating them, granulation unmistakably occurred at the common boundaries of the crystals, where strain can logically be expected.

That metal after solidification is usually in a strained condition is without doubt. The strains existing in the solid metal at high temperature may be quite sufficient to cause recrystallization. In this case it is easy to believe that recrystallization (or granulation) may take place.

The presence of infinitely small particles of mechanically held inclusions is perhaps another cause of granulation. The presence of inclusions disturbs the equalized atomic forces in the metals. Such a condition would result in the unequal distribution of atomic forces and the natural tendency would be for the metal to remedy the situation at the first opportunity.

Fig. 21—Macrostructure of Steel Containing an Excess of Phosphorus. Etched with Nitric Acid. 4x.

In photomicrograph, Fig. 11 is shown an eutectoid structure of a fully grown dendrite. The structure is a mass of grains, each grain radiating from an inclusion situated at its center. This observation has been made by the author on many other steels, but in this particular sample, the central position of the inclusion is quite obvious.

Dendrites and Grain Size

From numerous observations made by the author, the grain size in steel castings does not depend upon the size of the original dendritic growths. The structure of cast low carbon steel is shown in Fig. 12. The dendritic skeletons, the structure of which is

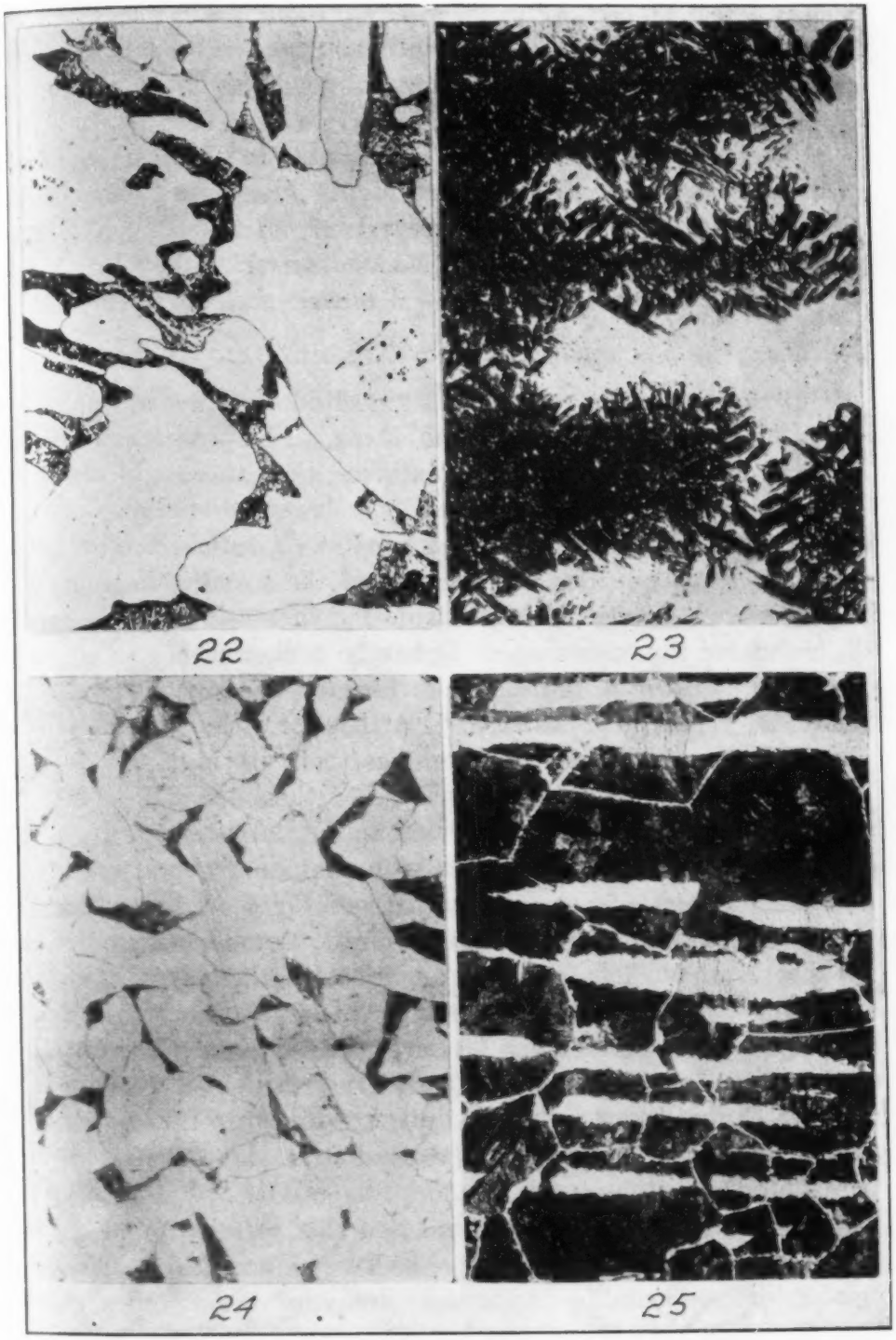

Fig. 22—Microstructure of Steel Shown in Fig. 21. Etched with Nitric Acid. 160x. Fig. 23—Microstructure of Steel Shown in Fig. 21. The Steel was Quenched from 1850 Degrees Fahr. in Water. Etched with Nitric Acid. 160x. Fig. 24—Microstructure of the Same Steel as in Fig. 21. Heated to 1900 Degrees Fahr., and Held for Twenty Hours. Etched with Nitric Acid. 160x. Fig. 25—"Banded" Structure in Steel. Etched with Nitric Acid. 50x.

shown in photomicrograph, Fig. 9, were detached from this steel. The similarity in structure is quite obvious, although the body of the casting was found to contain dendrites much smaller in size than the dendrite skeletons.

The photomicrographs of Figs. 13 and 14 illustrate the same point, but were taken from steel higher in carbon. One of the photomicrographs (Fig. 13) was taken from the body of a steel casting possessing rather small dendrites, while the other (Fig. 14) represents the structure of a dendritic skeleton of large size.

Rate of Solidification and Non-metallic Inclusions

If the size of the grains in cast steel does not depend upon the size of the original dendrites and if, as we know, we can control the size of the grains by heat treatment, then there appears to be no reason why we should concern ourselves with the size of dendrites and pay attention to the *rate of solidification* of our ingots.

But let us consider the question of the so-called non-metallic inclusions which are held mechanically in steel, and which are visible under the microscope. It has been shown that such inclusions are of utmost importance whenever the question of "fatigue" in metals is considered. Furthermore, the distribution of impurities in steels is as important as their amount.

Fig. 15 shows the microstructure of the unetched surface of ordinary steel with the normal amount of impurities. It is the same steel as shown in Fig. 8, but the photomicrograph (Fig. 15) was taken from a spot quite far from the spot shown in the sketch. If we examine the surface of the sample represented by the sketch, we shall find that the non-metallic inclusions are often found in patterns such as that produced in photomicrograph Fig. 16. Such a pattern can be developed only when there is full opportunity for unobstructed dendritic growth, because the inclusions, having lower melting points, are forced to follow the dendritic formation. The size of the dendrites, where the inclusions were found, is rather large.

It would be erroneous to think that this pattern found on the sample quite near the shrinkage cavity is due to the increased amount of non-metallic impurities: many of the samples examined show but a slight increase in this sort of inclusion. In photomicrograph Fig. 17 is shown the unetched surface of a dendritic skeleton. The complete abscence of inclusions should be noted.

In places, however, these inclusions are decidedly segregated as is shown in photomicrographs Figs. 18 and 19, the latter showing, in addition, the etched structure of a dendritic branch.

A piece of steel of very large dendritic structure was subjected to heat treatment such as to bring about a very fine structure (Photomicrograph, Fig. 20). This heat treatment could not, however, obliterate the excessive segregation of non-metallic impurities, which appear to form a continuous film. No heat treatment can improve steel showing inclusions in continuous patterns. In most cases it will fail under mechanical tests.

Rate of Solidification and Segregation

Certain elements, which do not form inclusions, but which are in steels in solid solution, have a tendency to segregate upon solidification. Elements such as silicon and manganese, when present in the usual amounts, can be diffused by heat treatment with relative ease. Carbon and phosphorus not only segregate, but their mutually repellent action results in pronounced local segregations of these two elements. The slow solidification of an ingot will naturally promote this action of separation and may lead to serious irregularities in the heat treatment of such steel ingots.

Let us first of all evaluate the true significance of a chemical analysis. A chemical analysis of an ingot gives us *the average* analysis of that volume of steel, represented by the drillings taken from it. But this analysis *does not* give us any idea as to the distribution of the elements, through even this volume. Consequently, while the analysis furnished to us by a chemist might well be within the specification limits, yet we have no right to say that in this steel we might not find small areas containing a certain element, far in excess of the permitted amount. On the other hand, there might be areas containing much less of the same element than the chemical analysis would lead us to believe. Now, this consideration becomes an important one when we begin to consider the behavior of carbon and phosphorus toward each other. As soon as the amount of phosphorus exceeds a certain percentage (the actual figures depend upon many considerations) the "divorce" action between carbon and phosphorus takes place. In certain cases, this action goes so far that the dendritic segregation in steels is revealed by ordinary etching with nitric acid, which blackens carbon bearing parts of steels. This can be seen in

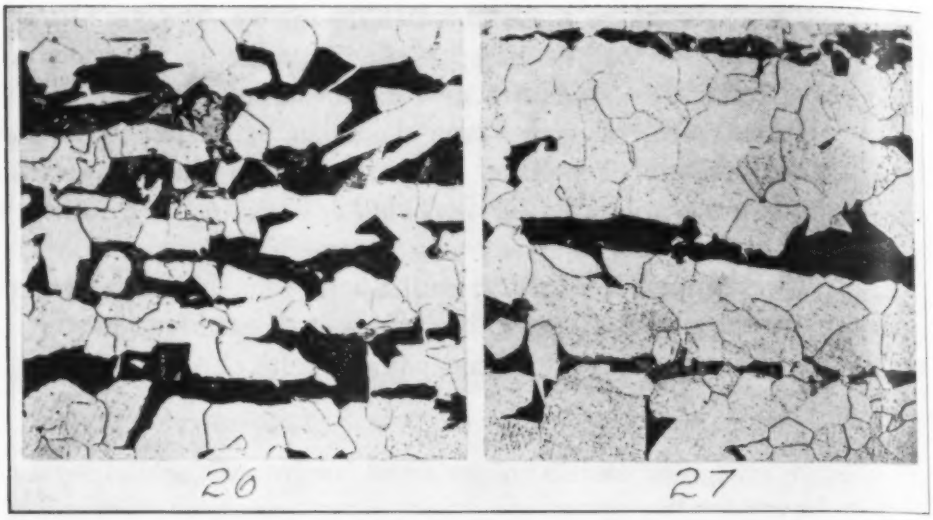

Fig. 26—"Banded" Structure Practically Eliminated by Quick Solidification of Steel. Etched with Nitric Acid. 100x. (W. P. Benter.) Fig. 27—"Banded" Structure in Steel Solidified Slowly. Etched with Nitric Acid. 100x. (W. P. Benter.)

Figs. 21 and 22. The last photomicrograph (Fig. 22) may indeed be taken for a low carbon cast steel and one is naturally led to ask whether these considerations mean anything at all since the steels are almost always subjected to some kind of heat treatment.

It was found, however, that the steel shown in two previous photomicrographs (Figs. 21 and 22) does not respond at all to the ordinary heat treatment. If kept for half an hour, at 1850 degrees Fahrenheit and quenched in water, it presents a peculiar structure (photomicrograph, Fig. 23), which is difficult to describe in metallographic terms. This structure shows, however, that the segregation still persists. Moreover it was found that it persisted when this steel was given another treatment of 5 hours at 1900 degrees Fahrenheit. The scratch test showed that the black areas were much harder than the rest of the sample. Only by heating this steel to 1900 degrees Fahrenheit and holding it at that temperature for over 20 hours was it possible to bring about the usual structure of annealed low carbon steel (photomicrograph, Fig. 24).

Often a hot-rolled bloom shows a distinct "banded" structure such as one shown in photomicrograph, Fig. 25.[2] This bloom was 14 inches square and was hot-rolled from an ingot of 32 x 32

[2]This and the three following photographs are taken from the thesis by W. P. Benter, entitled "Ingot Conditions as a Cause of Banding in Steel After Hot Working," and submitted to the Department of Metallurgy, Carnegie Institute of Technology for the degree of Metallurgical Engineer. The writer wishes to thank Mr. Benter for permission to use his photographs and results.

inches, having the following composition:

	per cent
Carbon	0.48
Manganese	0.87
Phosphorus	0.016
Sulphur	0.024
Silicon	0.20
Vanadium	0.19

After solidification but before becoming cold, according to information received by the author, the ingot was placed in a soaking pit and heated for rolling. The finishing temperature was about 1850 degrees Fahrenheit and immediately after finishing, the bloom was buried in hot ashes. At the end of five days, the bloom was removed from the ashes and its temperature was found to be about 500 degrees Fahrenheit.

In order to verify that the banded structure as shown had its origin in the conditions of solidification and was brought about, first, by undisturbed and excessive segregation of certain elements, in all probability, phosphorus and carbon, and second, by the "divorce" action of the two named elements, the following experiment was performed:[3]

> ".....The steel used was that which is called the S. A. E. grade and had the following range of analyses:
>
> | C | Mn | Ph | S | Ni | Cr |
> |---|---|---|---|---|---|
> | 0.10-0.20 | 0.30-0.60 | 0.040 | 0.045 | 1.00-1.50 | 0.45-0.75 |
>
> Three heats were made, all within the above analysis range. These heats were made in the open-hearth with all conditions as nearly the same as was possible in the practical operation. They were poured at a low temperature in order to insure more rapid solidification, in the ingot molds.
>
> "The only radical difference in practice in the three heats was the fact that heats Nos. 1 and 2 were poured into cold molds while heat No. 3 was poured into molds which were warm. The term "cold" and "warm" are merely relative terms. The cold molds had not been used for 24 hours and were at a temperature of about 150 degrees Fahrenheit. The "warm" molds were at a temperature of about 800 degrees Fahrenheit. The molds were all of the same size. After pouring they were held for 12 hours before being charged into soaking pits in order to be sure beyond doubt that the

[3]Quoted from a thesis by W. P. Benter.

centers were absolutely solid. They were then rolled into blooms 6 inches square.

"Representative blooms were selected from each heat for examination. Specimens were cut from the blooms at a position midway between the center and outside of the bloom in each case. These specimens were rough cylinders about 1 inch in diameter and 2 inches long. They were heated to 1750 degrees Fahrenheit, held at this temperature for one hour and allowed to cool slowly in the furnace over night.

"The specimens then were examined under the microscope on planes parallel to the direction of rolling of the blooms. The specimens from heats No. 1 and No. 2 were found to contain indistinct bands of ferrite very much broken up by pearlitic grains. The specimen representing heat No. 3, however, contained wide bands of ferrite in most cases entirely free from pearlite."

The structural conditions of steel after the heat treatment as described by Mr. Benter are illustrated in photomicrographs, Figs. 26 and 27.

The experimental evidence obtained in this laboratory and in the actual conditions of steel making and the conclusions drawn from it are, it seems, worthy of the steel-maker's attention. We may have a normal amount of impurities distributed equally through the steel. This amount may be harmless. We may have the same amount of the same impurities either segregated in one place (those that form solid solutions) or forming continuous patterns at times of very intricate form (inclusions). Such a condition may be very harmful. The two are the direct result of the manner of solidification. The evidence seems to be that a very pronounced segregation as well as the large and intricate patterns of inclusions are mostly found in steels, which show a decidedly large dendritic structure. Large dendritic structure is an evidence of large dendrites formed in steels during very slow solidification.

Dendritic segregation is by no means the sole cause of defective steel and a pronounced dendritic structure is not a sufficient reason for the rejection of steel. When the dendritic segregation is composed of elements mutually repellent, or when its presence is manifested by continuous segregations of non-metallic

impurities, steel, as a rule, will be found to be defective. This condition cannot be remedied by heat treatment and must be prevented by suitable control of the process of solidification. It is believed emphatically that the process of solidification has not received the proper amount of attention.

The Author

Vsevolod Nicholas Krivobok was born in Russia, November, 1892. After the completion of his studies at the Politechnical Institute of Petrograd he was employed at the Putiloff Works in Petrograd in the pressing and forging department. The Government of Russia sent him to America as a junior member of the Artillery Commission in 1915-1917, and he was transferred to Russian Mission of Ways of Communications, 1917-1919. In 1919 he returned to Russia and was assigned the duties of an engineer in the metallurgical department, ministry of railways, Government of South Russia.

V. N. KRIVOBOK

In 1920 he returned to the United States and completed his graduate work in 1924, at Harvard University. He was research associate in the bureau of metallurgical research, at Carnegie Institute of Technology, Pittsburgh, in 1924, and is at present associate professor of metallurgy of that institution.

ELECTRIC ANNEALING OF MAGNETIC MATERIALS FOR TELEPHONE APPARATUS

By W. A. Timm

Abstract

This paper briefly describes the annealing equipment used by the Western Electric Company at the Clinton Street shops up to 1909 and later at the Hawthorne Plant, in contrast to the more recently installed electrical equipment. The first named equipment was of the coal fired type and the second named equipment was of the oil-fired type. In 1924 one electric recuperative furnace was installed at the Hawthorne Plant to replace the oil-fired manual control furnaces. These furnaces are described in considerable detail. Numerous illustrations are included showing the furnaces, transformer room and methods of packing and handling the annealing pots.

General

In the transmission of speech by telephone, the various connections of the talking, signalling and supervising circuits established between subscribers are made by electromagnetic means. To connect the telephones of two subscribers a large number of relays must operate in succession at predetermined times.

Additional electromagnetic devices are used for signalling and for keeping direct, alternating, pulsating and voice currents limited to their respective channels.

As indicated by these general statements, the proper operation of the telephone system depends upon the exactness of the operation of these electromagnetic devices, which, in turn, depend in a great measure upon the quality and heat treatment of the magnetic materials used in the construction of these pieces of apparatus.

The telephone equipment used by the Bell Telephone System is manufactured by the Western Electric Company, Incorporated, in its plants at Hawthorne, Illinois, and Kearny, New Jersey.

Amount of Materials Requiring Annealing

The magnitude of the annealing problem at the Western Electric Company perhaps can be realized when it is stated that about

The author, W. A. Timm, is associated with the Western Electric Company, Chicago.

eight thousand pieces of appartus of regular manufacture are used to perform some electromagnetic function. When these pieces of apparatus are resolved into their magnetic parts, it is found that a net load of about six tons of magnetic parts must be annealed daily. While this weight is relatively small compared with the annealing programs of some other factories and foundries, it should be noted that magnetic parts for telephonic uses are small, having an average weight of a few ounces. In addition to this, certain parts require special methods of packing to prevent distortion during annealing.

In addition to the parts to be used for strictly magnetic purposes as mentioned above, about five tons of small parts are annealed daily to facilitate subsequent machining operations.

Equipment Used by the Western Electric Company for Annealing Parts

The annealing of materials for telephone uses began with the first steps of the manufacture of the telephone itself. The first annealing equipment used by the Western Electric Company was installed in the old Clinton Street shops in Chicago about forty years ago. This consisted of one small charcoal-fired furnace, which was replaced a number of years later by several coal-fired furnaces. These coal-fired furnaces were used in the Clinton Street shops until 1909. At that time the Hawthorne Plant took over all of the annealing work and a battery of ten oil-fired furnaces were introduced for that purpose. Changes were made at times to improve the operation of these furnaces, and in 1921 six furnaces of the same type were added to meet output requirements.

Some of these furnaces are still in use for parts which are annealed to facilitate subsequent machining operations only. All parts annealed for magnetic purposes are now handled in an electric recuperative annealing furnace which will be described later.

Description of the Oil-Fired Furnace

The battery of the oil-fired furnaces of the type which handled the entire product from 1909 to 1924 is shown in Fig. 1. These furnaces are single door, batch-type furnaces having inside dimensions of 7 feet, 6 inches long, 26 inches wide and 18 inches high. They were heated by a single oil burner located at the front of the

furnace, and the temperature of the charge was controlled by the manual adjustment of this burner. In this figure the operators are shown in the operation of loading the furnace with annealing pots. Three operators were required to perform this operation; one to operate the air hoist; one to place the pot in the clamping fixture and the third to slide the pot into the furnace.

This loading operation brings the operators into close contact with the hot furnace. The unloading operation is a still more un-

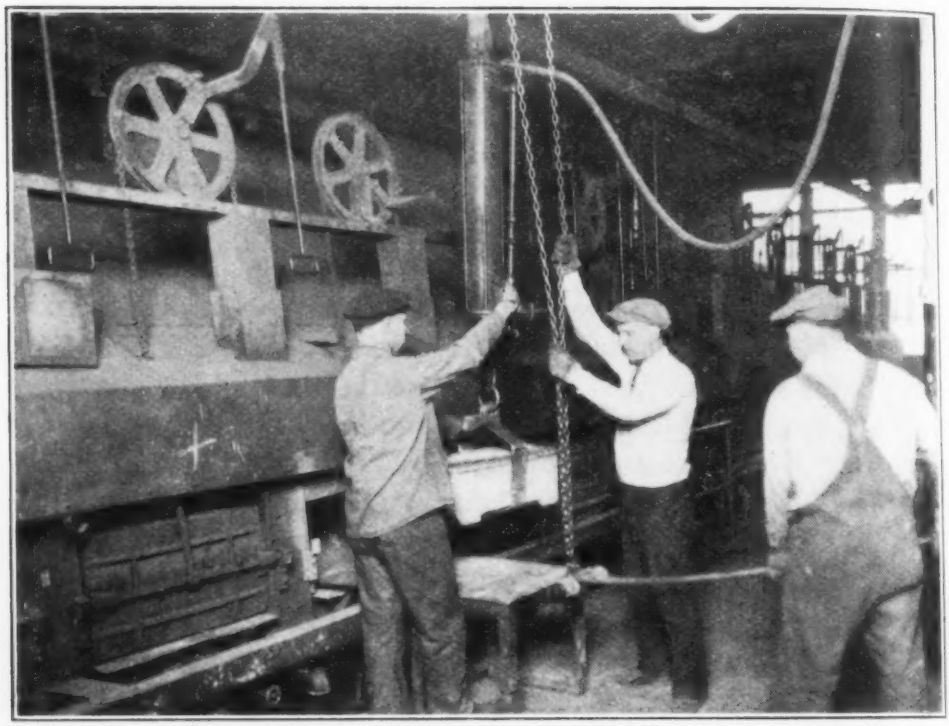

Fig. 1—Battery of Oil-Fired Furnaces of the Type which Handled the Entire Product from 1909 to 1921.

comfortable job, inasmuch as the hot pots are removed from the furnace by this method. These hot pots are placed on trucks and moved to some other corner of the room where they cool and in turn heat the room. The open furnaces and the removal of the hot pots often make working conditions extremely uncomfortable during the summer months.

In addition to the unsatisfactory operating features, the temperatures of these oil-fired furnaces could not be controlled with the accuracy required by the more severe magnetic requirements which were being placed on the apparatus. Tests on these furnaces

had shown variations as much as 150 degrees Fahr. in the temperatures of the charges in the various pots. It was also found that the heating conditions of the furnaces were constantly changing, probably due to the change of the direction of the flame within the furnace.

These two factors, the manual control of only a single burner

Fig. 2—Old and New Types of Annealing Pots Used in the Annealing of Parts.

for each furnace and the day by day variations of the interior conditions of the furnace made it quite impossible to obtain a uniformly annealed product from these furnaces.

The type of annealing pot used with these furnaces is shown in the upper part of Fig. 2. The parts to be annealed were placed in these pots and covered with cast-iron chips which were shaken down to fill the spaces between the parts. The sheet metal cover

shown in the picture was then placed over the parts and was luted to the sides of the pot with fireclay. A layer of chips was then placed over the plate.

This method of packing the pots to prevent excessive scaling of the parts was not particularly satisfactory, since after the anneal-

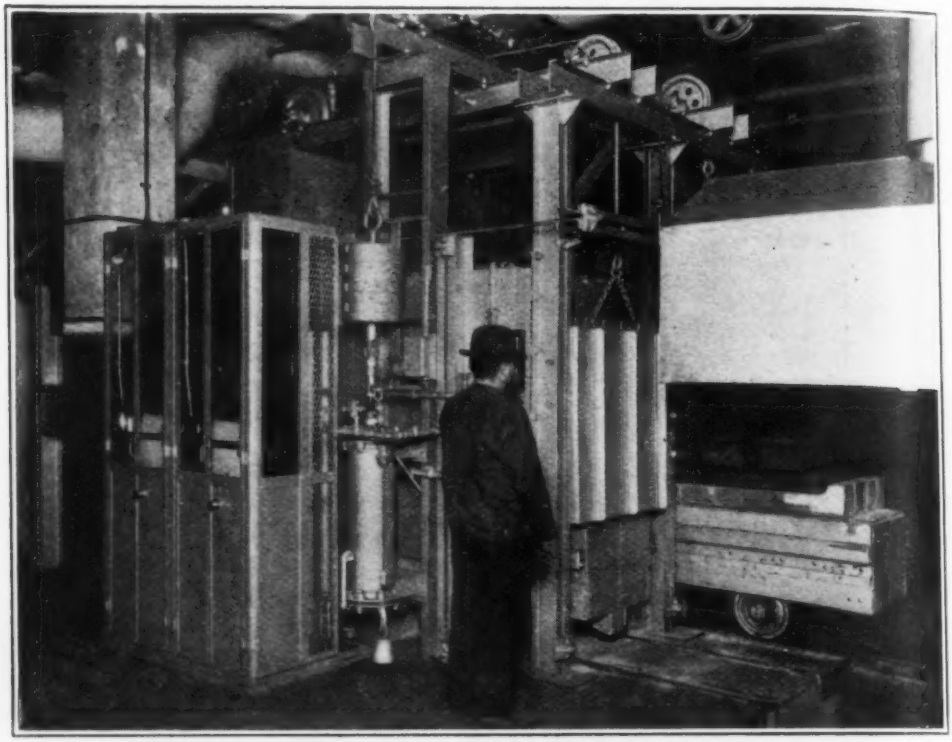

Fig. 3—Car-Type Recuperative Electric Furnace Showing Operator at Control Valves which Raise and Lower the Doors and Moves the Cars Through the Furnace.

ing operation the chips had to be screened from the parts. This was a very dusty and disagreeable operation. Furthermore, in many cases the iron chips stuck to the parts requiring an expensive hand brushing operation to remove them.

Purchase of new Annealing Furnace

The description of the oil-fired furnaces has been given to furnish a picture of the conditions which brought about the purchase of an electric recuperative furnace. This furnace was purchased in 1924 in an effort to improve the magnetic quality of the magnetic materials used in telephone apparatus and to increase the uniformity of the product, which was also of great importance. At that time increased production demands made necessary the

purchase of additional equipment. Accordingly, steps were taken to find the best furnace the market offered which would meet the particular needs of our product. The product of practically all the furnace suppliers was studied. In some instances actual shop trials of the furnace equipment were made in other plants. The

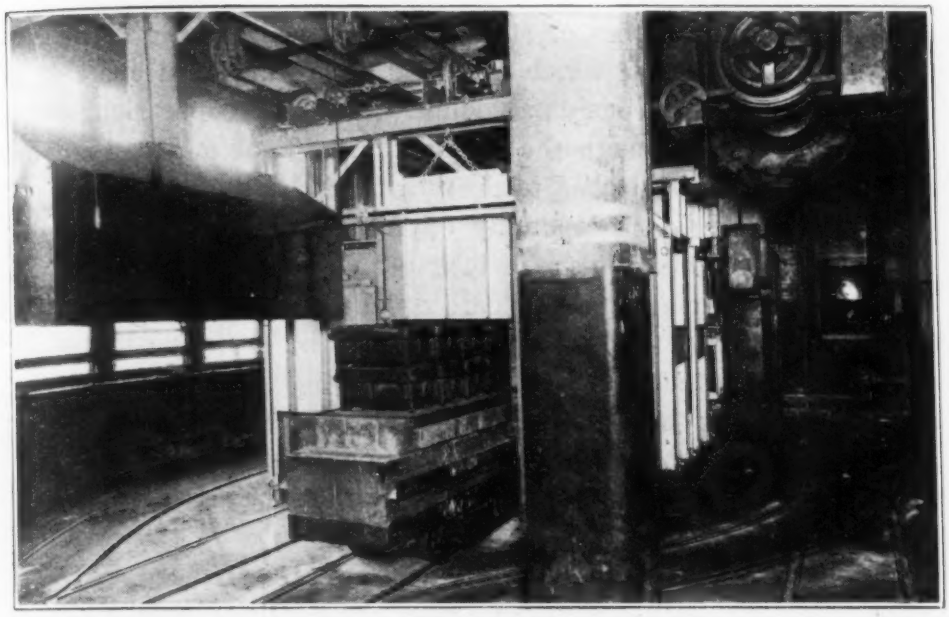

Fig. 4—Another View of the Car-Type Recuperative Electric Furnace as Shown in Fig. 3.

furnace was selected which we believed would best be able to meet our exacting magnetic requirements and which greatly improve the handling of the work before and after the annealing operations. After careful study of all propositions submitted, the car-type recuperative electric furnace shown in Figs. 3 and 4 was selected.

Before considering the chief points of interest in the construction and operation of this furnace, it may be stated that, on account of the relatively low ceiling height of about twelve feet and other limitations of this particular location, the furnace had to be modified somewhat to suit our particular needs. We had just sufficient room to install the furnace with none to spare.

Description of the Electric Annealing Furnace

A description of the furnace can probably best be begun by mentioning the car-type feature. It will be noted from Fig. 4 that the annealing pots containing the annealed parts are placed upon

a furnace car. The gross weight of the eight pots which are placed on the car is somewhat over three thousand pounds. The weight of an empty pot is about equal to the weight of the parts it contains. These pots are moved through the furnaces on the cars which, in addition to transporting them, has the added function of providing heat to them.

In Fig. 3 the operator standing at the side of the furnace is operating an air valve which raises all of the doors on one side of the furnace simultaneously and, subsequent to this operation, moves the furnace cars through the furnace. The door and car moving mechanisms are interlocked to prevent them from operating except in proper sequence. Diagonally opposite from the control valve alongside of which the operator is standing is a similar valve which controls the movement of the cars on the track nearest the operator. By having the air control valves in these positions the operator has a direct view of the car entering the furnace.

As shown in the diagram Fig. 5, two tracks pass through the furnace from end to end. The cars on one track move in a direction opposite to the cars on the second track. This movement of the cars is essential to the recuperation feature of this furnace. The interior of the furnace is divided into three zones. First a preheating and cooling chamber; then the heating chambers; and third another preheating and cooling chamber. At the ends of these chambers are doors which are raised and lowered pneumatically by the air control valves previously mentioned.

Since the cars and the pots of material are moving in opposite directions and since the heating chambers are at the center of the furnace, a cold charge enters the preheating chamber on one track, while on the other track a hot charge enters it from the heating chamber. In this manner a hot and cold charge are placed side by side under the arch of the preheating chamber, and the heat from the hot charge is transferred to the cold charge.

Fig. 6 shows in the foreground the preheating and cooling chamber. The single arched roof over both tracks is to be noted. The bottom ends of the raised doors between the preheating chamber and the heating chambers can also be seen. The heating chambers are directly beyond these doors. In one heating chamber is shown a car with its load of annealing pots, while in the other chamber an empty car has been placed, so that the heating elements on both

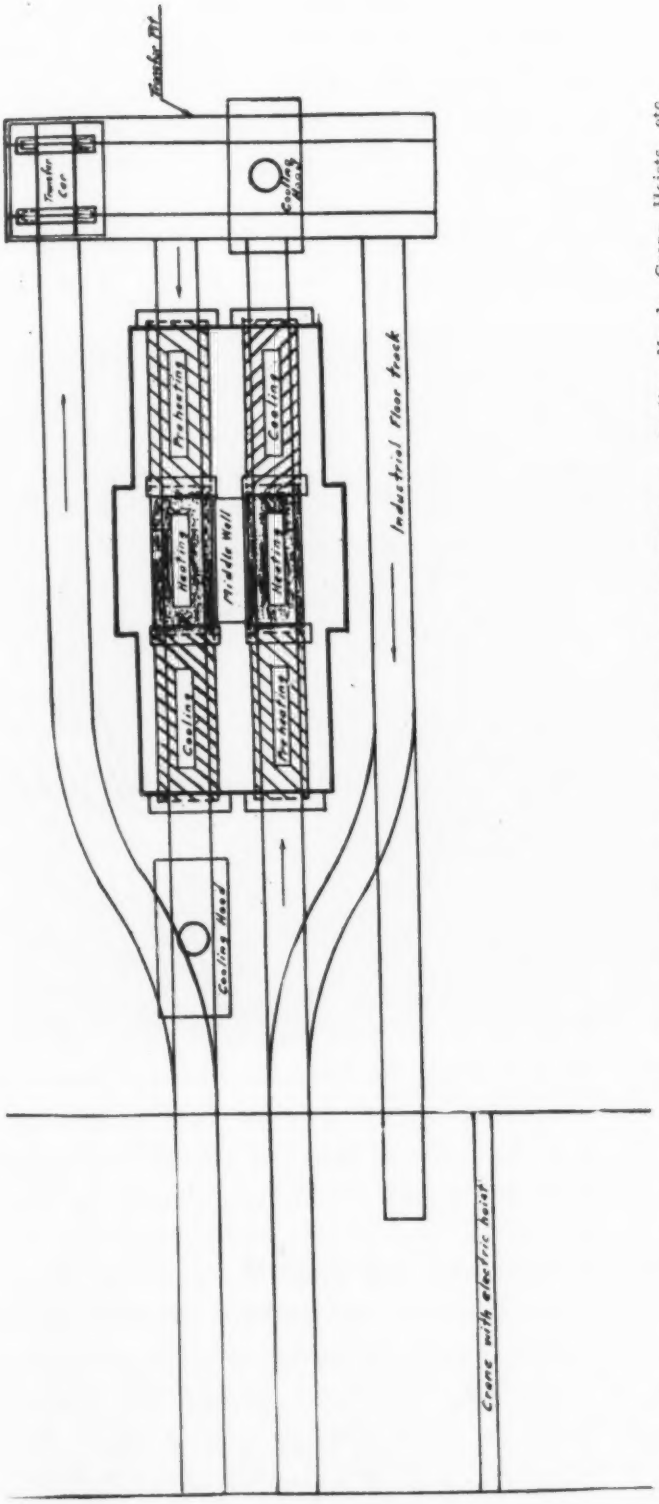

Fig. 5—Diagram of the Arrangement of the Recuperative Electric Furnaces Showing Trackage, Cooling Hoods, Crane Hoists, etc.

the side walls and the roof arch can be seen. It is to be noted that the two heating chambers are entirely separated by a common wall on which heating elements are placed. Looking through the heating chamber the preheating and cooling chamber at the other end of the furnace can be seen. This chamber is identical with the one in the foreground.

The position of the thermocouples for the temperature controls is near the heating elements at the side walls. The thicker tubes

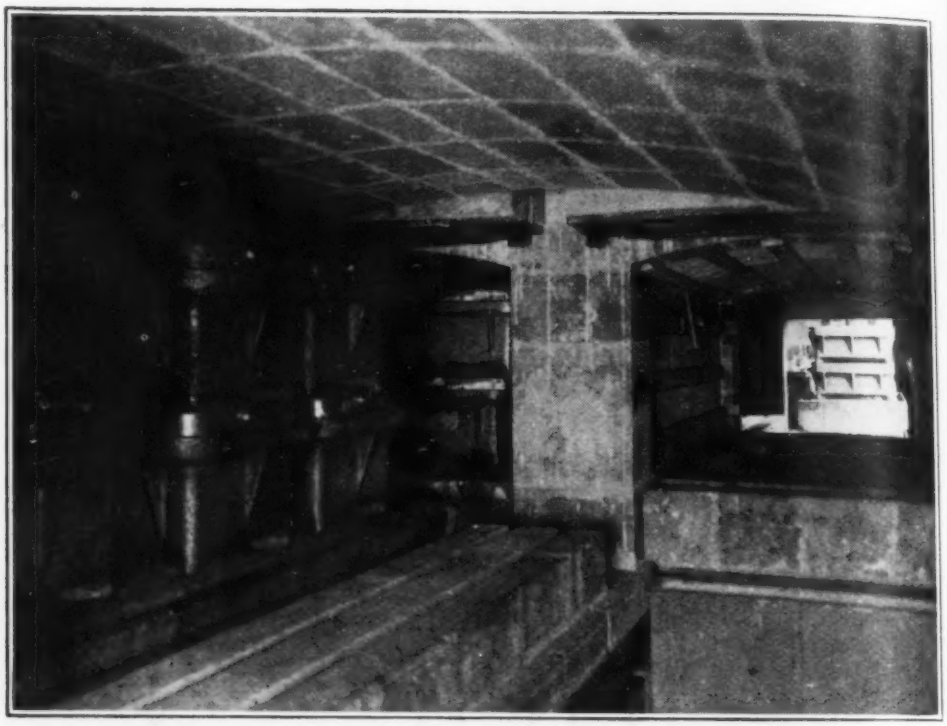

Fig. 6—View Taken from Within the Recuperative Electric Furnaces Showing the Arrangement of Pots and Electric Resistors.

extending down into the heating chamber are the temperature cut-outs, whose function is to prevent the possibility of overheating the furnace.

By the aid of Fig. 5 we will trace the cycle of the annealing operation through the furnace. A furnace car with its charge of eight pots at room temperature is moved into the preheating chamber. Immediately afterward a car on the opposite track is moved into this same chamber. This charge coming from the heating chamber has a temperature of about 1650 degrees Fahr. After remaining side by side for about six hours the initially cold charge

has reached a temperature of about 700 degrees Fahr. while the other charge has cooled about the same amount. At the end of the six-hour period the cooling charge is moved from the furnace while the heating charge is moved into the heating chamber. Since the time-temperature cycle of the cars on both tracks is the same, two charges each preheated to about 700 degrees Fahr. enter the heating chambers at the same time. Here they remain for another six hours until they reach a predetermined temperature, which we will assume to be 1650 degrees Fahr. After this period each charge is moved from the heating chamber to the cooling chamber where it gives up about one half of its heat to the next charge entering the furnace.

In this manner each charge entering the heating chamber has already been preheated to about 700 degrees Fahr. while each charge leaving the furnace has already been cooled to about that same temperature by doing useful work in heating the next charge.

Before leaving the consideration of the movement of the furnace cars, we will note from Figs. 3 and 4 that outside of the furnace at either end an exhaust hood is suspended by cables. In practice when a car having a temperature of about 700 degrees Fahr. is romoved from the furnace, it is placed under one of the cooling hoods which is pulled down over the stack of pots to the car top. This hood is provided with a motor-driven fan which exhausts the heated air arising from the pots to the exterior of the building. After the charge has remained beneath the hood for about five hours, its temperature has been reduced to less than 200 degrees Fahr. and the pots can then be handled without difficulty.

In summing up the time cycle used in annealing the charges just considered, we find that the charge was first obtaining a preheat in the preheating chamber for six hours, that it remained in the heating chamber for six hours before being moved to the second preheating chamber, where it was cooled for another six hours, and finally stood outside of the furnace under the cooling hood for a period of five hours. This made a total of twenty-three hours to complete the annealing operation.

Power Input to Electric Annealing Furnace

Approximately 158 kilowatts of power at a pressure of 440 volts, and 3 phase connection is supplied for these electric furnaces. This power is delivered through a 250,000 circular mil under-

ground cable to the transformer room shown in Fig. 7. Here three single-phase transformers, arranged in delta connection, transform the power to 84 volts and 42 volts, at which pressure it is distributed to the heating elements of the furnace.

The location of this transformer vault is directly below the east end of the furnace. This position reduces the length of the

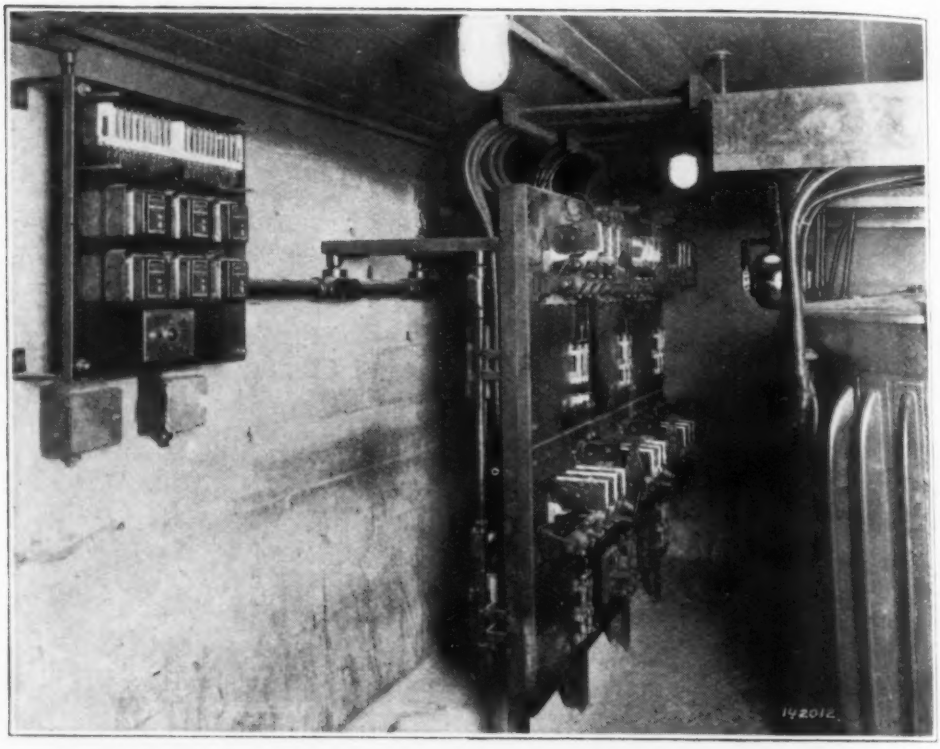

Fig. 7—Transformers and Control Panels for Controlling the Furnace Temperatures.

secondary leads and the resulting line losses to a minimum. In Fig. 7, the control panels for controlling the furnace temperatures can be seen directly opposite the transformers.

The power delivered to the furnace is divided equally between the elements in two heating chambers. The two preheating chambers are not equipped with heating elements, although the design of these chambers is such that heating elements can be installed if desired.

The power delivered to each heating chamber (about 79 kilowatts) is divided in approximately three equal parts. Somewhat more than one-third is delivered to the heating elements, located in the top of the furnace car directly below the heat resisting alloy

car top plate. The power for these heating elements is delivered to the car through three contact shoes located at the right side of the car between the wheels. Three brushes located at the bottom of each heating chamber make contact with these shoes when the car is in place. The remaining two-thirds of the power is divided about equally between the heating elements on the side walls and roof of the furnace. Some of the elements of both the side walls and the roof form each of these two circuits. By this distribution of the power, each circuit is kept a balanced three-phase load at all times.

In heating a charge of material, full power is delivered to all of the elements of the car top, side walls and roof of the furnace. This amount of power is delivered until the furnace reaches the temperature desired. When the furnace reaches this temperature, all the power which is being delivered to the two, side wall and roof circuits is cut off automatically. The car top heating elements are connected to a separate temperature control circuit. When the furnace temperature at the car top reaches this predetermined value, the power is cut off from this circuit also. When the temperature falls slightly, these connections are again made automatically. In this manner the temperature of the furnace is accurately controlled.

The temperature recording instruments may be seen in Fig. 3 mounted in the cabinet at the left of the operator. As mentioned before, the control panels on which are mounted the control relays and fuses are located in the transformer vault. Since these relays are quite noisy in operation their location in the isolated transformer vault materially reduces the noise in the annealing room.

Auxiliary Equipment

The parts to be annealed in the electric annealing furnace are placed in heat resisting alloy annealing pots. The general design of this pot is shown as the lower pot in Fig. 2. Parts are placed in the pot without any packing material as shown in Fig. 8. After the parts are placed in the annealing pot the groove around the top is filled with foundry loam, mixed with water to form a stiff mud and the cover is put on. A tongue on the cover embeds into the loam. After the cover is in place the loam shown on the outside is smoothed with a trowel.

Eight annealing pots are then placed upon the furnace car by means of the crane and hoist shown in Figs. 9 and 10. The time

required to load or to unload a car is about three minutes. At this point it may be added that the entire furnace charge can be moved ahead one step in less than two minutes. If necessary, a furnace car could be loaded and placed in the furnace in less than five minutes.

After a furnace car is loaded, it is pushed by the operators over the track shown in Fig. 5, and placed directly in front on the

Fig 8—Showing the Method of Packing the Pots Without Any Packing Material.

entrance to the preheating chamber. At the front end of the furnace the movement of the cars is made over the track and switches indicated, while at the rear of the furnace a transfer car is used.

Since this type of transfer car requires a pit about fifteen inches deep, its use is thereby prohibited at the front of the furnace where the floor space must be utilized for loading purposes.

Improvement in Working Conditions Made Possible by the Electric Annealing Furnace

The improvement of the electric recuperative furnace over the oil-fired furnaces previously described may be summarized as follows:

1. The working conditions are greatly improved.

2. With the electric furnace no hot pots need be handled by the operators and no manual lifting is required.

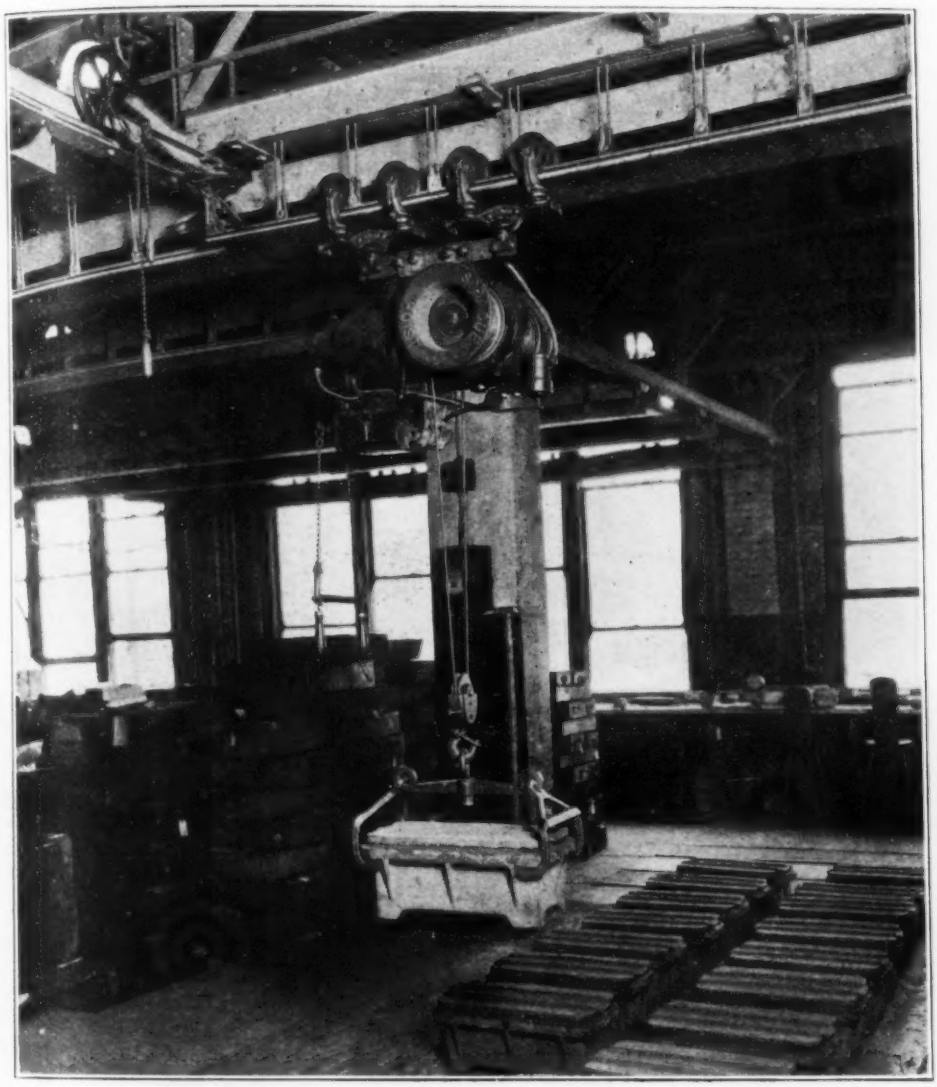

Fig. 9—Electric Crane Used in Handling the Annealing Pots.

3. No hot pots are stored in the annealing room as the cooling hoods exhaust the heat from the building.

4. The electric furnace equipment does not produce any gases of combustion or fumes to contaminate the air; a condition sometimes present when the oil-fired furnaces are operated.

5. There are no excessive or discomforting room temperatures

near the furnace, because of the high degree of insulation of the electric furnace.

6. The operation of the electric furnace is noiseless, while the oil-fired furnaces required a noisy air-blast.

Economy of Operation of the Electric Annealing Furnace

The electric furnace has been found to be more economical to operate than the oil-fired furnaces. Its operation during the year

Fig. 10—Electric Crane Used in Loading the Annealing Pots on the Furnace Car.

1925 resulted in an appreciable saving. These economies were realized on the following basis.

a. The recuperative principle of the electric furnace utilizes the waste heat of the hot pots to heat other pots.

b. One operator can charge into and discharge from the electric furnace in less than three minutes, two car loads of material equivalent to the capacity of four oil-fired furnaces. In the oil-fired furnaces this work would take three operators twenty minutes to perform.

c. No constant attention is required to operate the electric furnace after the temperature control apparatus is set, while in the operation of oil-fired furnaces one operator's entire attention is required to regulate oil and air valves.

d. In the electric furnace larger annealing pots which are packed without chips permit greater weight of parts to be packed in each pot. An appreciable saving has been realized in the reduction of the labor of packing and unpacking pots. Since no chips are used in the pots placed in the electric annealing furnace, the annealed parts do not need to be cleaned by brushing and filing.

e. When the sixteen oil-fired furnaces were operated, the total repair charges amounted to about $3000.00 per year. The annual repairs on the electric annealing furnace which has the capacity of about eleven oil-fired annealing furnaces has been quite nominal.

f. In addition to the economies given above, the electric annealing furnace offers new and greater opportunities for developing cheaper and more satisfactory methods of annealing.

Improvement in Quality of Annealed Product

Greater temperature uniformity and control is possible in the electric annealing furnace because of automatic temperature control features. These devices reduce the maximum temperature differences to be found in any of the eight annealing pots of a charge to about 30 degrees Fahr.

In the operation of the oil-fired furnaces the quality of the annealed parts were sometimes affected by changes in the quality or condition of the fuel oil. This condition is eliminated in the electric furnace.

No furnace gases are present in the electric furnace to contaminate the product.

The following is a comparison of the magnetic characteristics of material annealed in the oil-fired furnaces and material annealed in the electric furnaces. All of these tests were made on the same lot of material, and are, therefore, strictly comparable.

The magnetic characteristics of magnetic material annealed in the oil-fired furnace, based on tests made on 15 samples are as follows:

Maximum Permeability		Maximum Induction Gausses		Remanence Gausses		Coercive Force Gil./cm.	
Min.	Max.	Min.	Max.	Min.	Max.	Min.	Max.
3525	4645	16,200	16,600	9100	10,550	0.715	1.30

The magnetic characteristics of similar samples from the same lot of magnetic material, but annealed in the electric furnace are as follows:

Maximum Permeability		Maximum Induction Gausses		Remanence Gausses		Coercive Force Gil./cm.	
Min.	Max.	Min.	Max.	Min.	Max.	Min.	Max.
4400	4900	16,200	16,400	9770	10,500	0.86	1.07

It is to be noted that the material annealed in the electric furnace has superior magnetic quality as well as greater uniformity than the material annealed in the oil-fired furnaces.

The above characteristics are of particular importance in the operation of telephone relays, which is one of the most essential pieces of apparatus of the telephone system.

From the time that the electric furnace just described has been used for annealing magnetic parts, there has been a marked improvement in the magnetic quality and the magnetic uniformity of apparatus made from this material. This condition greatly aids the subsequent manufacturing operations and contributes to the improvement of the efficiency of the telephone system in general.

Conclusion

In the above brief description of the annealing equipment used by the Western Electric Company, Incorporated, it has been possible to do but little more than point out the improvements which have resulted from the use of the electric annealing furnace. For the particular needs of this company in annealing of large quantities of small parts to meet rigid magnetic requirements, it has been demonstrated that the car-type recuperative annealing furnace is well adapted to this work.

The Author

W. A. Timm was born April 14, 1889, at Bay City, Michigan. He received his early education in the grammar schools of Bay City and was graduated from the

Bay City high school in 1907. He then attended the University of Michigan, graduating in 1912, with the degree of Bachelor of Science in Electrical Engineering.

Following his college work Mr. Timm associated himself with the Western Electric Company, Inc. in August, 1912. From August, 1912, to September, 1913, his time was occupied with the student course given by this company. From September, 1913, to May, 1923, he was chiefly concerned with engineering problems dealing with manufacturing methods, working on the improvement of methods and equipment in connection with the following types of apparatus:—loading coils, repeating coils, retardation coils, auto transformers, condensers, black enameled wire, miniature lamps, vacuum tubes, electric furnaces, electric soldering irons, electric heating devices, permanent magnets, heat treatment of magnetic materials. At the present time Mr. Timm is chief of magnetic materials heat treatment equipment department in charge of the improvement, by development or purchase, of furnace equipment used for annealing, hot forming and magnet hardening processes.

W. A. TIMM

Educational Section

These Articles Have Been Selected Primarily For Their Educational And Informational Character As Distinguished From Reports Of Investigations And Research

THE CONSTITUTION OF STEEL AND CAST IRON
PART V

By F. T. Sisco

Abstract

The present installment, the fifth of the series, takes up the structure of high carbon steels, and discusses the structural changes in these steels when they are heated and cooled through the transformation ranges. The whole discussion of the constitution of steel is summarized. Methods are given for the calculation of the amount of the various structural constituents present in hypoeutectoid, eutectoid, and hypereutectoid steels.

IN the previous installment we discussed the changes taking place in steels containing 0.85 per cent or less carbon when cooled slowly from a high temperature. In the present installment we will take up the changes occurring in steels containing more than 0.85 per cent carbon cooled in the same way. That completed we will summarize the whole briefly.

It is very important that we become thoroughly familiar with the transformations occurring in slowly cooled steels and with the structural changes taking place at these various temperatures. A knowledge of what goes on in steel when it is heated or cooled through these transformation points helps us to understand the underlying principles of heat treatment.

As steel treaters we are all familiar with the operations necessary in normalizing, annealing, hardening and tempering. We all know that when we heat a piece of high carbon tool steel to 1400 degrees Fahr. and quench it in water or oil it becomes hard. This knowledge that steel hardens when treated in this way is of inestimable value, but it becomes of still more value if we know why it becomes hard. The knowledge gained through a study of

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the constitutional changes taking place at elevated temperatures helps us to understand the *why*.

Heat treatment is made possible by the fact that steel undergoes structural changes when heated or cooled through various temperature ranges. By taking advantage of the structural changes which occur when the steel is heated through these transformation points we are able to control the resulting mechanical properties of the metal and fit it to our needs.

Structure of High Carbon Steel

On a previous page we learned that steel is composed of iron and iron carbide, in other words, ferrite and cementite. Ferrite and cementite under conditions of stable equilibrium form an aggregate known as pearlite which contains 0.85 per cent carbon or 12.5 per cent cementite and 87.5 per cent ferrite. Steels containing less than 0.85 per cent carbon contain less than 12.5 per cent cementite. As the eutectoid ratio is always 12.5 per cent cementite to 87.5 per cent ferrite it follows that in the steels containing less than 0.85 per cent carbon the structure will consist of pearlite and free ferrite.

Now, if a steel contains more than 0.85 per cent carbon it will contain more than 12.5 per cent cementite. As the eutectoid ratio is always the same these high carbon steels will consist of pearlite and free cementite.

The accompanying table[65] gives the percentages of the constituents present in the high carbon steels when in stable equilibrium.

Carbon Per Cent	Total Cementite Per Cent	Total Ferrite Per Cent	Pearlite Per Cent	Free Cementite Per Cent
.90	13.5	86.5	99.7	0.3
1.00	15.0	85.0	98.0	2.0
1.10	16.5	83.5	96.3	3.7
1.20	18.0	82.0	94.6	5.4
1.30	19.5	80.5	92.9	7.1
1.40	21.0	79.0	91.2	8.8
1.50	22.5	77.5	89.5	11.0
1.60	24.0	76.0	87.8	12.2
1.70	25.5	74.5	86.1	13.9

As the carbon content increases above 0.85 per cent the free cementite increases in increments of about 1.7 per cent for each

[65]Sauveur, Metallography and Heat Treatment of Iron and Steel, 1926, page 63.

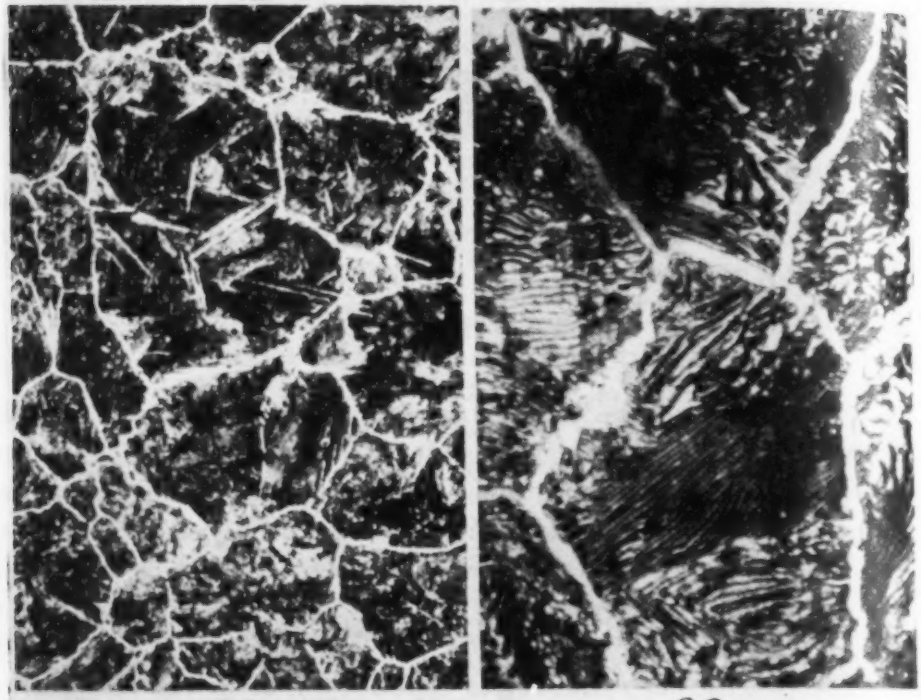

Fig. 32—Photomicrograph of Steel Containing 1.20 per cent Carbon. Structure is Pearlite (dark) and Cementite (light) at Grain Boundaries and in the Pearlite Grains. Etched with Alcoholic Nitric Acid. Mag. 100 X. Fig. 33—Photomicrograph of Steel Containing 1.43 per cent Carbon. Structure is Pearlite (dark laminated) and Cementite (light) at Grain Boundaries. Mag. 500 X.

0.10 per cent carbon as shown in the above table. In the previous installment[66] we saw that in hypoeutectoid steels the free ferrite increased in increments of about 12 per cent for each decrease of 0.10 per cent carbon. For example, a steel containing 0.70 per cent carbon, 0.15 per cent below the eutectoid point has a structure consisting of 81.9 per cent pearlite and 18.1 per cent ferrite. A steel containing 1.00 per cent carbon, 0.15 per cent above the eutectoid point, contains 98 per cent pearlite and only 2 per cent free cementite.

From the percentages given in the above table we can readily conceive what the structure of high carbon steels will be.

A hypereutectoid steel containing 1.20 per cent carbon will contain 94.6 per cent pearlite and 5.4 per cent free cementite. The structure will consist of pearlite and a small amount of cementite.

[66]See table, page 474, TRANSACTIONS, Vol. 10, September, 1926.

Fig. 32 shows the structure of a 1.20 per cent carbon steel at 100 diameters. The dark grains are pearlite. The cementite appears as a network at the grain boundaries and in a few of the grains as white needles. This structure is characteristic of hypereutectoid steel, cooled slowly from above the transformation range.

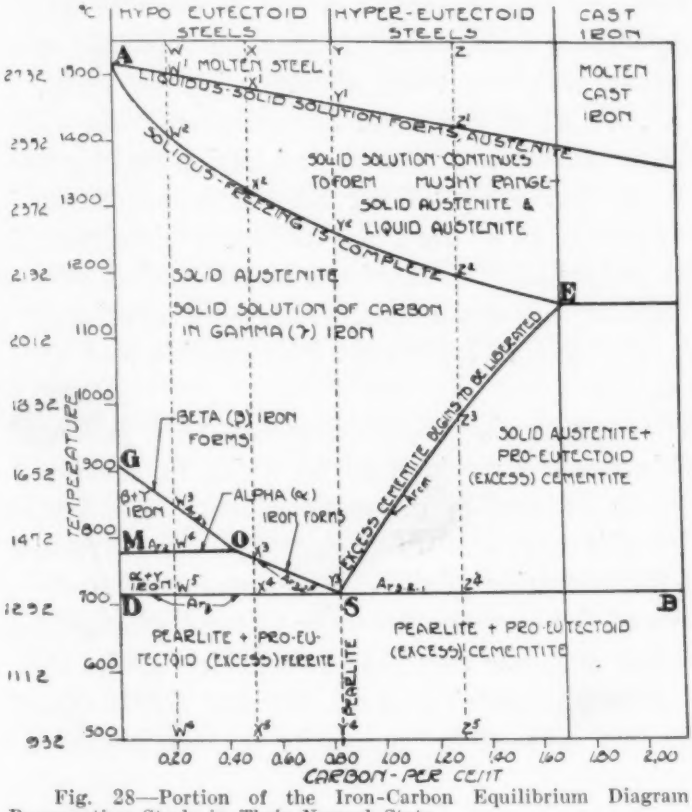
HYPO EUTECTOID STEELS
HYPER-EUTECTOID STEELS
CAST IRON
MOLTEN STEEL
LIQUIDUS-SOLID SOLUTION FORMS AUSTENITE
MOLTEN CAST IRON
SOLID SOLUTION CONTINUES TO FORM MUSHY RANGE-SOLID AUSTENITE & LIQUID AUSTENITE
SOLIDUS-FREEZING IS COMPLETE
SOLID AUSTENITE
SOLID SOLUTION OF CARBON IN GAMMA (γ) IRON
BETA (β) IRON FORMS
ALPHA (α) IRON FORMS
EXCESS CEMENTITE BEGINS TO BE LIBERATED
SOLID AUSTENITE + PRO-EUTECTOID (EXCESS) CEMENTITE
PEARLITE + PRO-EUTECTOID (EXCESS) FERRITE
PEARLITE + PRO-EUTECTOID (EXCESS) CEMENTITE
PEARLITE
TEMPERATURE
CARBON - PER CENT

Fig. 28—Portion of the Iron-Carbon Equilibrium Diagram Representing Steels in Their Normal State.

Fig. 33 shows the structure of a steel containing 1.43 per cent carbon at 500 diameters.[67] The laminations of the pearlite grains are readily apparent. The white cementite here forms a heavy network at the grain boundaries.

Constitutional Changes in High Carbon Steels

In Fig. 28 which has been reproduced here, we will follow the solidification and cooling of a 1.25 per cent carbon steel. This steel is represented by the ordinate Z. At 2800 degrees Fahr. (1538 degrees Cent.) the metal is entirely molten and consists of a solution of 1.25 per cent carbon (18.7 per cent iron carbide,

[67] From Sauveur, Loc. Cit., page 59.

Fe_3C) in molten iron. The metal cools unchanged until a temperature of 2625 degrees Fahr. (1440 degrees Cent.) is reached. Here the ordinate Z intersects the liquidus A. At this point, Z^1, solidification begins by the deposition of a solid solution of iron carbide (cementite) in iron. The first solid solution that solidifies contains less carbon than the average of 1.25 per cent. As the temperature falls, more of the solid solution, of increasing carbon content, solidifies until at 2190 degrees Fahr. (1200 degrees Cent.), Z^2, all of the metal has solidified. As soon as solidification starts, diffusion of the carbon begins and by the time the point Z^3 is reached the solidified mass is homogeneous in carbon content.[68] From Z^2 to Z^3 the metal undergoes no allotropic change. During this temperature interval the metal consists of a solid solution of cementite in gamma iron. This solid solution or austenite is made up of about 18.7 per cent iron carbide and 81.3 per cent gamma iron.

At 1775 degrees Fahr. (925 degrees Cent.) a transformation occurs. This would be evidenced by a jog in a cooling curve if one were made on steel of this composition. The transformation occurring at this temperature (point Z^3, Fig. 28) marks the beginning of the liberation of the excess cementite, that is the cementite in excess of the amount necessary to form the aggregate pearlite. A steel of 1.25 per cent carbon contains about 18.7 per cent cementite. Pearlite contains about 12.5 per cent cementite, hence there is 6.2 per cent excess cementite in solution in the gamma iron.

We have already noted that at 2100 degrees Fahr. (1150 degrees Cent.) about 25 per cent cementite is soluble in gamma iron. At 1300 degrees Fahr. (720 degrees Cent.) only 12.5 per cent cementite is soluble in gamma iron, hence as the temperature falls from 2100 to 1300 degrees Fahr. (along the line ES, Fig. 28) the cementite is expelled or liberated from solution.

In the case of our steel containing 1.25 per cent carbon or 18.7 per cent cementite; at any point above 1775 degrees Fahr. (975 degrees Cent.) the austenite is undersaturated, hence there is no change between the point Z^2 and Z^3. At the temperature 1775 degrees Fahr., 18.7 per cent cementite exactly saturates the austenite. Below this temperature the solid solution of cementite

[68] Except for a small amount of segregation, which may be omitted for the present.

in gamma iron is supersaturated and cementite is thrown out of solution.

Now, as the steel cools from 1775 (Z^3) to 1300 degrees Fahr. ($Z,^4$ Fig. 28) the excess cementite is being constantly expelled until at 1300 degrees Fahr. (720 degrees Cent.) the steel is now of eutectoid composition. At this temperature the Ar_1 transformation occurs and the austenite now of eutectoid composition is transformed into pearlite with evolution of heat.

From 1300 degrees Fahr. the steel, now made up of 93.8 per cent pearlite and 6.2 per cent excess (free or proeutectoid) cementite, cools unchanged to atmospheric temperature.

The temperatures representing the beginning of the expulsion of the cementite (line ES, Fig. 28) are known collectively as the Ar_{cm} transformation. The lower transformation occurring in high carbon steels and represented by the line SB where the austenite of eutectoid composition is transformed to pearlite is known as $Ar_{3\text{-}2\text{-}1}$ because at point S the lines GOS ($Ar_{3\text{-}2}$) and the line DS, (Ar_1), have merged into one line (denoted as $Ar_{3\text{-}2\text{-}1}$).

These two transformations Ar_{cm} and $Ar_{3\text{-}2\text{-}1}$ are the ones taking place in cooling. When steel is heated the corresponding transformations $Ac_{3\text{-}2\text{-}1}$ and Ac_{cm} occur at slightly higher temperatures.

In later installments we shall see that the line Ar_{cm} (and Ac_{cm}) is important as it fixes the normalizing temperatures of high carbon steel while the line $Ar_{3\text{-}2\text{-}1}$ ($Ac_{3\text{-}2\text{-}1}$) fixes the annealing and hardening temperature.

Summary

In order to fix firmly the facts brought out in our discussion of constitutional changes in steels when heated or cooled through various temperatures we will summarize these facts briefly.

1. The transformation points (also called critical points) in steel are those temperatures at which internal structural changes take place in the metal. At the transformation point an evolution of heat in cooling and an absorption of heat in heating occurs which may be detected by thermal analysis. A transformation (or critical) point is indicated by a jog in the heating or cooling curve.

2. The upper transformation point known as Ar_3 in cooling and Ar_2 in heating occurs in pure iron and in steels containing

less than 0.40 per cent carbon. The upper transformation point indicates the temperature of the change of gamma iron to beta (?) iron in cooling (Ar_3) or the change from beta iron to gamma iron in heating (Ac_3).

The structural change is from the face-centered cubic lattice of the gamma iron crystals to the body-centered cubic lattice of the beta (alpha) iron crystals. In pure iron Ar_3 occurs at about 1650 degrees Fahr. (900 degrees Cent.). The presence of carbon lowers the upper transformation point from 1650 degrees Fahr. for a pure iron to 1400 degrees Fahr. for a 0.40 per cent carbon steel. When a low carbon steel undergoes a transformation at Ar_3 the change is that of gamma iron free from carbon to beta iron free from carbon. The carbon present in the steel is thus concentrated in the gamma iron as yet untransformed.

3. The middle transformation point known as Ar_2 in cooling or Ac_2 in heating occurs in pure iron and in steels containing less than 0.40 per cent carbon. The middle transformation point indicates the change of the beta allotropic form free from carbon to the alpha allotropic form also free from carbon. This Ar_2 point occurs at the constant temperature 1400 degrees Fahr. (760 degrees Cent.) for pure iron and for all steels from 0.01 to 0.40 per cent carbon.

There is no evidence of a structural change at the Ar_2 point although iron not magnetic above Ar_2 becomes magnetic below this temperature. The Ar_2 point in pure iron and low carbon steels, although known definitely to exist, is still in dispute so far as its importance is concerned.

4. The Ar_{3-2} transformation is a merging of the upper, Ar_3 transformation and the middle, Ar_2, transformation and occurs in steels containing between 0.40 and 0.85 per cent carbon. The structural change at Ar_{3-2} is the change from carbon-free gamma iron to carbon-free alpha iron (either directly, or instantaneously through beta iron). The increasing carbon content in the steel causes a lowering of the Ar_{3-2} transformation from a temperature of 1400 degrees Fahr. until it merges with the Ar_1 line at 1290 degrees Fahr. As the allotropic change involves the practically carbon-free gamma to practically carbon-free alpha, the carbon (cementite) in the steel is concentrated in the gamma iron as yet untransformed. This concentration continues until the remaining

austenite contains 0.85 per cent carbon which occurs when the temperature has fallen to 1290 degrees Fahr.

5. The lower transformation Ar_1, does not appear at all in pure iron. In very low carbon steels it is very faint, but increases in intensity as the carbon content increases until it reaches a maximum with 0.85 per cent carbon. The transformation Ar_1 occurs at the constant temperature 1290 degrees Fahr. (700 degrees Cent.) for all carbon percentages. The same transformation in heating, Ac_1, occurs some 50 degrees Fahr. higher. The Ar_1 transformation marks the change of the austenite (gamma iron containing carbon in solid solution) of eutectoid composition to the aggregate pearlite. Obviously its intensity depends upon the amount of carbon.

6. The upper transformation in hypereutectoid steels, Ar_{cm} in cooling, marks the beginning of the precipitation of the excess cementite from the saturated solid solution of cementite in gamma iron.

The temperature at which the Ar_{cm} transformation occurs depends upon the amount of cementite. With the maximum amount of excess cementite in solution (12.5 per cent) the beginning of the transformation occurs at 2100 degrees Fahr. When 6 per cent excess cementite is in solution the transformation begins at about 1775 degrees Fahr. and when the eutectoid ratio is reached and no excess cementite is in solution the transformation point has disappeared.

7. A low carbon steel of about 0.20 per cent carbon will have three transformation points, at Ar_3, carbon-free gamma iron changes to beta, at Ar_2 the beta changes to alpha and at Ar_1 the remaining austenite now of eutectoid composition changes to pearlite.

A medium carbon steel (about 0.50 per cent carbon) will have two transformation points, at $Ar_{3\text{-}2}$ the carbon-free gamma iron changes to alpha iron and at Ar_1 the remaining austenite now of eutectoid composition changes to pearlite.

A steel containing about 0.85 per cent carbon has one transformation point $Ar_{3\text{-}2\text{-}1}$ where the austenite of eutectoid composition is transformed into pearlite. A high carbon steel of more than 0.85 per cent carbon has two transformation points, Ar_{cm} which marks the beginning of the precipitation of the excess

cementite, and $Ar_{3\text{-}2\text{-}1}$ where the austenite of eutectoid composition is transformed into pearlite.

It should be remembered that the lower transformation point Ac_1 (or $Ar_{3\text{-}2\text{-}1}$) marks the change of the austenite or gamma iron containing about 0.85 per cent carbon (as cementite) in solid solution, to the aggregate pearlite. The upper transformations in a steel containing less than 0.85 per cent carbon marks an allotropic change in the ferrite (pure iron) relatively free from carbon. In the steels containing more than 0.85 per cent carbon, the upper transformation marks the beginning of the precipitation of the excess cementite. These are the fundamentals in the constitution of the steels.

Calculating the Structural Composition of Steel

It is sometimes very convenient to be able to determine the proximate composition of a sample of steel. The ultimate composition, that is, the percentage of carbon as determined by chemical analysis, does not tell us what the constitution of that steel may be. Knowing that a piece of steel contains 1.25 per cent carbon gives no clue as to how much is pearlite, how much is cementite and how much of this cementite is combined or free. The method of calculating the structural composition of any steel in its normal condition is worked out in full in Professor Sauveur's well known text, "The Metallography and Heat Treatment of Iron and Steel", to which the reader is referred for a complete discussion.

In calculating the structural composition of steel we have three general classes of steel to deal with:

(1) Hypoeutectoid; steels containing less than 0.84 per cent carbon;[69] (2) eutectoid; steels containing about 0.84 per cent carbon; and (3) hypereutectoid; steels containing more than 0.84 per cent carbon.

As eutectoid steels are made up of one metallographic constituent only we will discuss them first. All of the carbon in steel is contained in the cementite. The formula weight of cementite (iron-carbide, Fe_3C) is 180. The atomic weight of carbon is 12. Then:

(1) Total cementite percentage = 15 × total carbon per-

[69] For convenience in calculating, pearlite is assumed to contain 0.84 per cent carbon. The correct carbon percentage lies between 0.83 and 0.88 per cent. Sauveur uses 0.85 per cent in his formulas.

centage.[70] Now, if the carbon content of the pearlite aggregate is 0.84, the percentage of cementite will be

15 × 0.84 per cent = 12.60 per cent, or in round numbers 12.5 per cent. Since the remainder of the pearlite aggregate is ferrite (practically carbon-free) the amount of ferrite is 87.5 per cent. It follows from this that the cementite-ferrite ratio in pearlite is 1 to 7.[71] Therefore 1 part of cementite is equal to 8 parts of pearlite or:

(2) Pearlite percentage = 8 × total cementite percentage.

Since the total cementite in the pearlite is 15 times the carbon percentage and the pearlite is 8 times the cementite, the pearlite will be

(3) Pearlite percentage = 120 × total carbon percentage.

Formula No. 3 gives a direct and simple method of calculating the amount of pearlite in hypoeutectoid and eutectoid steels from the carbon content. Hypoeutectoid steels contain pearlite and free ferrite; hypereutectoid steels contain pearlite and free cementite.

In the case of hypoeutectoid steels containing less than 0.84 per cent carbon we have the total ferrite in two structural conditions. We have first the ferrite of the pearlite, that is, the ferrite associated with the cementite in the pearlite aggregate and second, the excess ferrite not contained in the pearlite aggregate. The first is usually described by the terms combined ferrite, or eutectoid ferrite. As we have already noted, the second is known as excess ferrite, free ferrite or proeutectoid ferrite.

The calculation of the amount of free ferrite is very simple. Hypoeutectoid steels in their normal state are made up of pearlite and ferrite. If we calculate the amount of pearlite by formula No. 3 and subtract this from 100 per cent the result is the excess ferrite. To find the amount of combined ferrite we calculate the total cementite by formula No. 1 and the amount of pearlite by formula No. 2 or 3. The difference is the combined ferrite.

By these simple arithmetical calculations we are able to work out the structure of any steel containing 0.84 per cent carbon or less.

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[70] 12:180 :: 1 : X. X=15.

[71] 12.5 : 87.5 :: 1:X. X=7.

Let us apply the calculations to 5 steels of varying composition.

CASE I.—0.20 per cent carbon steel.

(1) Total cementite = 15 × 0.20 per cent.

Total cementite = 3 per cent.

(3) Pearlite = 120 × 0.20 per cent.

Pearlite = 24 per cent.

Free ferrite + pearlite = 100 per cent.

Then: 100 − 24 = 76 per cent free ferrite.

Now:

Pearlite − total cementite = combined ferrite.

24 − 3 = 21 per cent combined ferrite.

The steel is made up of

	Cementite	3 per cent
	Combined Ferrite	21 per cent
	Free Ferrite	76 per cent
or	Pearlite	24 per cent
	Free Ferrite	76 per cent
or	Cementite	3 per cent
	Total Ferrite	97 per cent

CASE II.—0.60 per cent carbon steel.

(1) Total cementite = 15 × 0.60 per cent.

Total cementite = 9.0 per cent.

(3) Pearlite = 120 × 0.60 per cent.

Pearlite = 72.0 per cent.

Free ferrite + pearlite = 100 per cent.

Then: 100 − 72 = 28 per cent free ferrite.

Pearlite − total cementite = combined ferrite.

or

72 − 9 = 63 per cent combined ferrite.

The steel is made up of

	Cementite	9 per cent
	Combined Ferrite	63 per cent
	Free Ferrite	28 per cent
or	Pearlite	72 per cent
	Free Ferrite	28 per cent
or	Cementite	9 per cent
	Total Ferrite	91 per cent

CASE III.—0.84 per cent carbon steel.

(1) Total cementite = 15 × 0.84 per cent.

Total cementite = 12.5 per cent (approximate).

(3) Pearlite = 120 × 0.84 per cent.

Pearlite = 100 per cent (approximate).

If the steel contains 100 per cent pearlite, no free ferrite is present.

Pearlite — total cementite = combined ferrite.

100 — 12.5 = 87.5 per cent combined ferrite.

The steel is made up of

Cementite	12.5	per cent
Combined Ferrite	87.5	per cent
Free Ferrite	0	per cent
or Pearlite	100	per cent

It should be understood that the simple method outlined above gives approximate results only. It is based on the pearlite-cementite relationship, that 1 part of cementite results in the formation of 8 parts of pearlite which contains 0.84 per cent carbon. Obviously this relationship is dependent upon the exact carbon content of the pearlite.[72]

In hypereutectoid steels the structure consists of pearlite and free cementite. The calculations in this case are a little more complicated than in the case of hypoeutectoid steels because the carbon is not all contained in the pearlite as in eutectoid and hypoeutectoid steel, but is divided between the cementite in the pearlite aggregate and the free cementite. Confronted by these difficulties we resort to the ferrite as a basis for the calculation because all of the ferrite is necessarily contained in the pearlite while the carbon is divided between the pearlite and the free cementite.

The first step is to calculate the total cementite present by formula 1.

(1) Total cementite percentage = 15 × total carbon percentage.

As the steel is made up of cementite (combined plus free) the ferrite can be found by subtracting the result of formula 1 from 100

100 per cent — total cementite = total ferrite.

[72] In "Metallography and Heat Treatment of Iron and Steel," Sauveur gives a more exact method for the calculation. See that text, pages 61 to 65, 1926 edition.

All this total ferrite is included in the pearlite. On a previous page we saw that 1 part of cementite plus 7 parts of ferrite gave 8 parts of pearlite. Then 7 parts of ferrite produce 8 parts of pearlite or

(4) 1 part of ferrite = 8/7 or 1.15 parts of pearlite (approximately).

We now have these two formulas (No. 1 and 4) to work from. Let us proceed to our illustrations.

CASE IV.—1.20 per cent carbon steel.

(1) Total cementite = 15 × 1.20 per cent.
Total cementite = 18.0 per cent.

Then: 100 per cent — 18 per cent = 82 per cent ferrite which is all contained in the pearlite.

Then: (4) 82 per cent × 1.15 = 94.3 per cent pearlite.

And: 100 — 94.3 = 5.7 per cent free cementite.

The structural composition of the steel is

	Free Cementite	5.7 per cent
	Pearlite	94.3 per cent
or	Total Cementite	18.0 per cent
	Ferrite	82.0 per cent
or	Combined Cementite	12.3 per cent
	Free Cementite	5.7 per cent
	Ferrite	82.0 per cent

CASE V.—1.50 per cent carbon steel.

(1) Total cementite = 15 × 1.50 per cent.
Total cementite = 22.5 per cent.

Then:

100 per cent — 22.5 per cent = 77.5 per cent ferrite which is all contained in the pearlite.

Then:

(4) 77.5 × 1.15 = 89.1 per cent pearlite.

And:

100 — 89.1 = 10.9 per cent free cementite.

The structural composition of the steel is

Free Cementite	10.9 per cent
Pearlite	89.1 per cent

or	Total Cementite	22.5 per cent
	Ferrite	77.5 per cent
or	Combined Cementite	11.6 per cent
	Free Cementite	10.9 per cent
	Ferrite	77.5 per cent

Sauveur[73] using the physical properties which he has assigned to the constituents ferrite, pearlite and cementite, gives a method of calculating tensile strength, ductility and Brinell hardness of hypoeutectoid, eutectoid and hypereutectoid steels. The values obtained by his calculations check very closely with actual test results of the various carbon steels in their normal, slow-cooled condition.

[73]Loc. Cit., pages 66-69.

Editor's Note—The next installment will discuss the effect of manganese, silicon, sulphur, phosphorus and other impurities on the structure of normal steels.

WEAR OF STEELS WITH PARTICULAR REFERENCE TO PLUG GAGES

By H. J. French and H. K. Herschman

(Continued from Page 717)

tical service. As a matter of fact any set of numerical values obtained on one job, would quite probably not apply to another even in the same plant. It is, therefore, felt that the author's selected procedure is best adapted to the work in view, but the importance of the surfaces gaged as well as the surface of the gage in contributing to wear should not be overlooked.

In conclusion, the authors are greatly pleased at the interest and discussion aroused by this paper and hope to extend their experiments much further.

Comment and Discussion

Papers and Articles Presented Before the Society and Published in Transactions Are Open to Comment and Criticism in This Column—Members Submitting Discussions Are Requested to give Their Names and Addresses

DISCUSSION OF H. B. KNOWLTON'S PAPER ON "FACTS AND PRINCIPLES CONCERNING STEEL AND HEAT TREATMENT—PART IV"*

By C. G. Williams†

In his reply to the discussion of his article (August issue of Transactions, page 305), Mr. H. B. Knowlton makes the assertion that the present writer neglected to mention steels with a manganese content of between 2 and 6 per cent.

It is very true, as Mr. Knowlton says, that most authorities state—as did Hadfield—that manganese steels having a manganese content between the limits given above are brittle, for such is the case when made with a carbon content such as those experimentors had to contend with, but it is by no means true that the same holds with steels of other limits of carbon content.

At the present time, outside of his own experiments, the writer can find out of a long list, only two authorities who quote for a higher manganese content than perhaps 1.75 per cent—Bullens[1] states: "Certain manganese steels with 1.50 to 2.78 per cent manganese, and a considerable carbon content, made in the electric furnace, have shown wonderful mechanical properties, and, in addition, will stand a tremendous amount of abuse in their thermal treatments without any great ill effects." And, speaking before the New York Chapter of American Society for Steel Treating and noted in the January 1926 Transactions J. H. Hall[2] stated: "A steel containing 9.50 per cent carbon and 2.50 per cent manganese has great possibilities for engineering service in the future. This steel, which has air hardening properties, can be made to give reduction of area and elongation of over 10 per cent by suitable heat treatment."

Along this same line, Sauveur[3] states: "The belief in the brittleness of pearlitic manganese steels is founded on Hadfield's statement that between 2 and 6 per cent of manganese, the steels are hopelessly brittle. On closer examination, however, it would seem as if this statement were true only in the case of rather high carbon steels cooled relatively quickly. Evidences have since been offered, notably by Guillet, showing that low carbon pearlitic manganese steels, slowly cooled, are not brittle", and further in a note: "Arnold reports that steel containing less than 0.10 per cent carbon and 5.50 per cent manganese

[1]Bullens—Steel and Its Heat Treatment, page 391.

[2]Transactions, American Society for Steel Treating, Vol. 9, January, 1926, page 158.

[3]Sauveur—Metallography and Heat Treatment of Steel, page 344.

*H. B. Knowlton, "Facts and Principles Concerning Steel and Heat Treatment—Part IV," Transactions, American Society for Steel Treating, Vol. IX, No. 1, page 111.

†The author of this discussion, C. G. Williams, is mechanical engineer, Sunstrand Engineering Company, Rockford, Ill.

has a tensile strength of some 145,000 pounds per square inch, and an elongation of 28.50 per cent.''

The writer has not felt that he was sufficiently known as a metallurgist to warrant rushing into the face of all the authorities who state that low manganese steels are brittle, with a statement to the contrary, but a close perusal of a considerable literature on this subject, has shown that nearly all or quite all of the statements regarding manganese steels are but quotations from Hadfield's writings, and are not arrived at through personal experiments. For that reason it would seem to be a case of giving Rover a bad name.

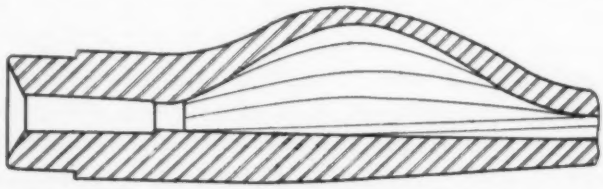
In the article in question, the writer stated: ''It is found that with a carbon content of from 0.20-0.30 per cent a manganese content of even 4.50 per cent will give a steel that when water-quenched at from 1255 to 1340 degrees Fahr. will give high tensile strength and ductility of about 25 to 30 per cent in two inches. The actual figures, with allowable variations for tests, are, carbon, 0.28 per cent; manganese, 4.45 per cent. Giving a tensile strength of 187,000 pounds per square inch, elastic limit of 162,000 pounds per square inch and an elongation of 26 per cent in 2 inches. Another steel of 0.21 per cent carbon, manganese 5.10 per cent gave a tensile strength of 142,000 pounds per square inch and an elongation of 32 per cent. The highest manganese content tested was 5.70 per cent; carbon 0.15 per cent, which gave a tensile strength of 151,000 pounds per square inch and an elongation of 26 per cent.

Another test of a steel with 0.22 per cent carbon; 4.80 per cent manganese, gave a tensile strength of 177,000 pounds per square inch and an elongation of 21 per cent. A steel with a carbon content of 0.08 per cent and a manganese content of 5.60 per cent gave a tensile strength of 158,000 pounds per square inch and an elongation of 32 per cent.

Others could be given, but it is not desired to prolong this discussion to a tiresome length.

All of the above tests were made on steels selected and made for a choice of some steel that would give a special long life to rifle barrels. These and many others were made up into rifle barrels and shot to destruction.

The last quoted steel gave a rifle barrel that after 65,000 shots had been fired from it, still had the riflings in very serviceable condition and the barrel still gave good accuracy. Pressures ran from 51,000 pounds per square inch



to 63,000 pounds per square inch, with no swellings nor stretchings of the barrel that could be detected. The pressures were slowly increased, and at last, at 155,000 pounds per square inch pressure, the barrel stretched as shown in the accompanying sketch but did not burst.

The writer has long felt that the automobile manufacturers were missing an especially useful steel when they overlooked a line of low-carbon, low manganese steels.

Reviews of Recent Patents

By
NELSON LITTELL, Patent Attorney
475 Fifth Ave., New York City
Member of A. S. S. T.

1,583,549, Method of Treating Aluminum-Silicon Alloys, Alfred George Cooper Gwyer and Henry Wilfred Lewis Phillips, of London, England, Assignors to the British Aluminum Company, Limited, of London, England.

This patent describes the method of treating aluminum-silicon alloys by adding to the molten alloy a peroxide of an alkali metal, for example sodium peroxide, which produces a strong exothermic reaction, cooling to the desired temperature and casting in a chill mold to accelerate the rate of cooling and produce an alloy having a greater dispersion of the silicon content and correspondingly improved physical properties. About 3 per cent of sodium peroxide is used in an aluminum silicon alloy containing 87 per cent aluminum and 13 per cent silicon.

1,587,992, Composition of Matter for Alloys, 1,587,993 Composition of Matter for Alloys, 1,587,994, Composition of Matter for Alloy of Metals, Ray L. Spitzley and Allen M. Thompson, of Detroit, Michigan, Assignors to Alloys Foundry Company, of Detroit, Michigan, a corporation of Michigan.

This series of patents relates particularly to an alloy of a composition capable of resisting a high temperature and oxidation, and particularly adapted for valves of internal combustion engines. The ingredients of the alloys are so proportioned as to produce a composition having a high melting point so that when the valve head is cast on the turned steel valve stem, the temperature of the alloy will be sufficient to integrally fuse the stem and head together, but will solidify before changing or distorting the enclosed section of the stem. The compositions may be varied according to various requirements. The following are given as typical examples:

	Per Cent
Nickel	56
Chromium	8
Manganese	7
Iron	26
Silicon	2
Tungsten	1

	Per Cent
Nickel	64
Manganese	7
Iron	14
Silicon	2
Tungsten	1
Copper	12

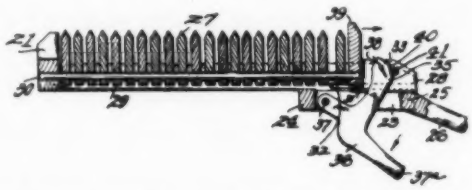
	Per Cent
Nickel	58
Chromium	8
Manganese	7
Iron	14
Tungsten	1
Copper	12

1,591,598, **Process of Producing Malleable Iron, Harry M. Williams and Alfred L. Boegehold, of Dayton, Ohio, Assignors to General Motors Research Corporation, of Dayton, Ohio, a corporation of Delaware.**

This patent relates to the process of reducing malleable iron which consists in incorporating into the iron a percentage of silicon higher than is ordinarily used in malleable iron practice and casting in a chill mold so as to chill rapidly and prevent the precipitation of flake carbon. The casting is then annealed from 1500 to 1700 degrees Fahr. for a period not exceeding 15 hours. The iron used preferably contains about 1.7 per cent carbon and upward of 2 per cent silicon.

1,600,266, **Clamp for Holding Knives, Frederick Armstrong, of Southbridge, Mass.**

This patent relates to a clamp for holding knife blades or the like while undergoing hardening, tempering, and other operations. The clamp



comprises end members 21 and 39, and intermediate clamping jaws 27, the jaws being slidably mounted on a rod 29 connected to one of the end clamping members. The knife blades are positioned between the clamping jaws 27 and the lever 36 actuated to move the jaws 27 together to firmly grip the base of the knife blades leaving the blade proper protruding in position to be heated or quenched or subjected to any other desired annealing and tempering operations.

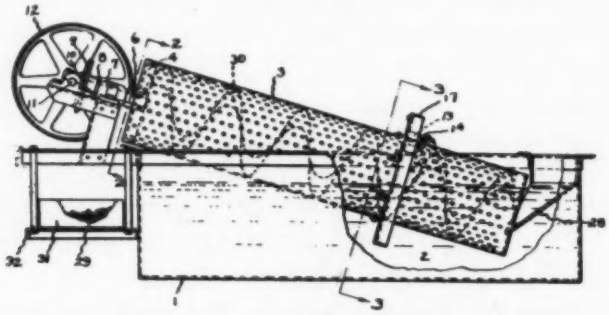
1,588,111, **Casting Having Silicon-Alloy Surface, Charles B. Jacobs, of Wilmington, Delaware, Assignor to E. I. Du Pont De Nemours & Company, of Wilmington, Delaware, a corporation of Delaware.**

This patent describes the method of producing steel or cast iron castings having a silicon alloy surface, which comprises casting the metal in ordinary green sand molds which have been lined or coated with a paste consisting of powdered ferro silicon and sodium silicate, which is applied to the mold surfaces and dried. The metal is poured into the mold at approximately the solidifying point thereof, so as to permit solidification of the major portion of the metal before the silicon has become dispersed throughout the body of the casting. The surface of the metal casting combines with the ferro silicon

coating of the mold to produce an alloy surface on the casting containing a high percentage of silicon.

1,601,497, Quenching or Washing Apparatus and the Like, Elmer G. Greene, of Cleveland, Ohio.

This patent describes an apparatus for quenching or washing metal parts. The apparatus comprises a tank 2 adapted to receive a quenching pickling or washing liquid and a rotating perforated drum 3 driven through a shaft 6 from the gears 9 and 10 and pulley 12. The lower edge of the



rotating drum is suspended on enclosed bearings 13 and 14 which are protected from contact with the solution in the tank 2. In the operation of the device, the articles to be quenched, washed or pickled, are dropped into the hopper 28 and are deflected into the lower end of the drum 3, the rotation of the drum and the location of the worm 30 therein elevating the articles from the liquid and discharging them into the receptacle 31 at the upper edge of the drum.

1,599,613, Recuperative Apparatus, Frank A. Fahrenwald, Cleveland Heights, Ohio.

Heated gases are delivered to the bottom of one chamber and discharged from the bottom of the other chamber of a pair of hollow upright recuperative chambers in this recuperator. Upwardly extending conduits of heat-conducting material are located in both chambers, and means are provided for causing an upward flow through all of these conduits of the gases to be heated.

1,599,010, Furnace-Wall Construction, Robert T. Haslam, Belmont, Massachusetts, Assignor to Standard Development Company.

Refractory, redundant exposures are positioned in a heating chamber in effective heat-radiating relation to an object to be heated and beyond the region in which combustion takes place. These redundant exposures abstract heat from the hot gases and radiate same upon the object.

1,600,724, Burner, William Henry Fitch, Allentown, Pa.

This pulverized fuel burner comprises a casing having an outlet opening in one end wall; a feed pipe projects through the other end wall and terminates within the casing. This feed pipe is in alignment with the outlet opening and is adapted to deliver a mixture of pulverized fuel and air. Means are provided for introducing an additional supply of air into the casing.

Abstracts of Technical Articles

Brief Reviews of Publications of Interest to Metallurgists and Steel Treaters

MELTING OF CAST IRON IN THE ELECTRIC FURNACE. By Oberingenieur E. Richards, Berlin. *Stahl und Eisen*, February 25, 1926, p. 249-254.

The principal disadvantages of melting in the cupola furnace are:

(1) the impossibility to manufacture an always uniform product,

(2) the impossibility to refine the cast iron sufficiently during melting,

(3) and to overheat it in a sufficient manner for the purpose of suitable casting.

The electric melting furnace serves only to refine the iron, melted in the cupola; its particular application is given by the possibility to diminish the contents of sulphur and phosphorus, by the manufacture of a uniformly composed cast iron and by the possibility of sufficient overheating.

In technical respect the basic electric furnace is superior to the acid furnace, but nevertheless this latter is often preferred to the basic furnace as the basic lining inclines to splintering off and to cracking.

To combine the advantages of the basic process with that of the acid trials were made with neutral refractory materials as carbon and chromium ore.

Then the author describes the "duplex" process, a combination of cupola-melting and refining in the electric furnace. The refining time depends on: (1) the temperature of the cupola furnace iron. (2) the desired degree of overheating of the final product. (3) the percentage of sulphur in the cupola iron. (4) the desired percentage of sulphur in the final product. (5) the amount of oxide and gas that must be removed. (6) and the greatness of the chemical treatment, being necessary.

The chemical reactions during the refining process and the melting progress are briefly outlined. Finally the manufacture of synthetic cast iron in the electric melting furnace is described.

Abstracted by Dr. Hans Pollack, Germany.

THE MANUFACTURE OF STEEL AND IRON WIRE IN GERMANY. By Dr. H. Altpeter. In *Wire*, October, 1926, page 194.

This article originally appeared in *Stahl und Eisen* and was translated by Carl Marx. In it the author sets forth the types of furnaces and lead baths used for the heat treatment of wire. The author states that the principal purpose for heat treating wire is to obtain a fine grained structure in the finished wire. He explains the two methods used—one "cementation" and the other "patenting." The results obtained are the same though the processes are different. Continuing, the writer describes these processes at some length

and illustrates the furnaces used for each. He points out the results obtained by the use of both methods and concludes his article by a discussion of lead baths and sand method of cooling and is of the opinion that scaling is prevented by the latter process.

RECRYSTALLIZATION OF IRON. By Dovrovidoff, A. N., and Grdina, Y. V. *Proceedings* of the Siberian Technological Institute (Tomsk), 1926, 1 (47), 34-46.

This is a report of the authors' study of the influence of different factors on the recrystallization of cold-worked iron. The present paper treats mainly of the degree of cold working, the temperature and the time of heating and the composition of the material. Soft iron was used in the experiments described containing 0.09-0.12 per cent carbon. Samples of it after careful annealing were submitted to stretching, the degree of which was measured by the residual elongation, which varied in the different experiments from 2 to 20 per cent. The stretched samples were then submitted to recrystallization at different temperatures and for different time intervals. The variations in the microstructure occasioned thereby were controlled by measuring the dimensions of the grains. From the data mentioned in this paper it follows, that the critical deformation of soft iron corresponds to an increase of length of 6-7 per cent; at greater as well as smaller elongation hardly any recrystallization could be observed. At the critical deformation of the soft iron the most convenient temperature lies at about 800 degrees Cent. Above and below this temperature a very pronounced retardation of recrystallization is observed. At these best conditions the recrystallization lasts the whole time of the heating process, its upper limit not having been attained by the authors even after 25 hours of heating. The recrystallization takes place sooner in iron containing carbon less than 0.1 per cent and as little as possible of other admixtures. In steel with 0.2 per cent carbon recrystallization is scarcely to be observed. *M. Oknoff and F. A. Miller, Leningrad, Russia.*

ON THE MECHANICAL PROPERTIES OF TITANIUM STEEL. By Kanzi Tamaru. Reprinted from the *Science Reports* of the Tohoku Imperial University, Series I, Vol. XV, No. 1, Sendai, Japan, April, 1926.

This paper deals with the mechanical properties of alloys, and states that the materials used in preparing these alloys were Swedish steels of varying carbon content and ferrotitanium. The author sets forth the chemical analyses of both these materials, and then proceeds to give the mechanical properties of carbon steels which in his opinion vary with their heat treatment. He then proceeds with his description of the making of titanium steel in a cryptol furnace and tells how oxidizing and nitrogenizing may be minimized, and sets forth the results obtained. This he further augments with graphs and tables. The author then gives consideration to titanium steel made in a vacuum furnace. A comparison between titanium and carbon steels is made by means of a graph. He is of the opinion that titanium steel melted in hydrogen has a much higher tensile strength than carbon steel; also that titanium steel has greater strength when melted in hydrogen than when melted in air; the yield point of titanium steel decreases with the carbon content; and that the contraction of area in titanium steels is much higher than that of carbon steels.

ON CAST-IRON. By Dr.-Ing. R. Hohage, Essen. *Kruppsche Monatshefte*, June, 1926, pages 101-109.

This paper deals with the structure and the changes in structure taking place in the heat treatment of cast iron alloys.

The iron-carbon alloys, containing more than 1.7 per cent carbon solidify either as white iron in which case the stucture consists of austenite-cementite eutectic, or as cementite and mixed crystals (pearlite) or in such a manner that the structure consists only of ferrite and graphite. Between these two types there are different structures influencing the properties of the high carbon-iron alloys to a high degree.

The purpose of the first series of this research was the investigation of the influence of varying percentages of silicon and manganese at a constant percentage of carbon.

The influence of the silicon content was examined on the following alloys:

Sample:	C	Si	Mn	P	S
A	3.10	0.59	0.62	0.59	0.11
B	3.25	1.12	0.70	0.47	0.11
C	3.23	1.65	0.71	0.54	0.13
D	3.30	2.02	0.67	0.64	0.13

The effect of an increasing amount of silicon under the same conditions on the graphite content and on the hardness may be seen in Table 1.

The percentage of carbon increases the graphite content very intensively

Table 1

Sample:	A	B	C	D
Silicon per cent	0.59	1.12	1.65	2.02
Graphite per cent	0.20	2.34	2.42	2.40
Brinell hardness	415	241	255	241

up to 1.1 per cent silicon; from there only in a minor degree. Consequently there exists a sudden transition from combined carbon to graphite. The correctness of this sudden change is proven by former investigations (*Journal* of the Iron and Steel Institute, 1906). These researches quoted the transition at a silicon content of 1.4 to 1.7 per cent. The difference from 1.1 to 1.4 or 1.7 per cent is probably founded on the higher carbon content of the present samples; the position of the change is probably fixed by the ratio of carbon to silicon. The Brinell hardness changes according to the increasing of the carbon content. The influence of the silicon may be distinctly seen in the structure. *Influence of Manganese:* the addition of manganese works against the formation of graphite; the hardness is increased. However, this is not exactly true when the amount of graphite is very high, that is in a cast iron with a high percentage of carbon and silicon; thereby the effect of manganese is compensated for.

The Brinell hardness throws light on the structure. When the Brinell number is greater than 400, the structure consists of austenite-cementite eutectic, pearlite and traces of graphite; from 400 to 280 Brinell the quantity of the austenite-cementite eutectic decreases and that of graphite increases; from 280 to 180 Brinell the structure consists of pearlite and graphite; below 180 Brinell appears free ferrite and from 140 to 120 Brinell the structure consists of ferrite-graphite and traces of pearlite. Knowing the chemical

composition, the thickness of the casting and by that the cooling conditions, the Brinell hardness makes it possible to predict the structure and by it the properties of the alloy.

Heat Treatment: The test pieces were annealed at 500, 600, 700, 830 and 900 degrees Cent. (932, 1112, 1292, 1526 and 1652 degrees Fahr.) for two days; the influence of the annealing may be distinctly seen in the structure. The sorbitic pearlite changes into a laminated one and this latter into a globular one. But not only the structure of the pearlite is influenced by the annealing treatment but the pearlite and the austenite-cementite eutectic are also decomposed under certain conditions; which are: the chemical composition, the rate of cooling and the annealing temperature.

Up to an annealing temperature of 600 degrees Cent. (1112 degrees Fahr.) the decrease of hardness is slight; at 700 degrees Cent. (1292 degrees Fahr.) and higher the hardness decreases; and it is at these points that the hardness numbers are most closely connected with the structure.

Abstracted by Dr. Hans Pollack, Germany.

CONTRIBUTIONS TO THE QUESTION OF ANNEALING OF BOILER IRON. By S. P. Nologdine. *Messenger of the metal Industry* (Moscow); 5-6 (1926); 5-9.

This paper deals with several examples of annealing of boiler iron at different temperatures and during different time intervals. On the parts of the metal, which were exposed in process of production to cold working a coarse-grained crystallization appears and an accompanying tendency to brittleness takes place. The annealing of boiler iron in the case of total or partial cold working may thus be liable to a starting of brittleness in it. To the author's opinion the annealing of such a material has necessarily to be accompanied by a control of its microstructure, with which it is easy to observe an abnormal recrystallization of the metal and prevent it with corresponding means.

M. Oknoff and F. Miller, Leningrad, Russia.

INDUSTRIAL FURNACES. By Charles Longenecker. In *Iron Trade Review*, September 30, 1926, page 843.

This is the second installment of a series on the design, construction and function of modern melting, heating and treating units and deals primarily with melting furnaces.

The author is of the opinion that melting furnaces are the heart of a steel company and that the open-hearth furnace is of greatest importance to the steel industry. He sets forth the improvement in port construction in recent years and devotes considerable space to a consideration of waste-heat and how to avoid it. He also discusses the various fuels used. The writer, by means of illustrations, presents views of the Ely puddling furnace; the Siemans-Martin open-hearth furnace fired with producer gas; a mechanical puddling furnace of the Roe type and the plan, back, side and front elevation of a straight-neck puddling furnace.

ON THE TERNARY DIAGRAM OF THE SYSTEM IRON, CARBON AND COPPER. By Torajiro Ishiwara, Takeo Yonekura and Toyozo Ishigaki. Reprinted from the *Science Reports* of the Tohoku Imperial University, Series I, Vol. XV, No. 1. Sendai, Japan, April, 1926.

The author sets forth that in Japan iron ores almost always contain small amounts of copper hence in that country the investigation of the effect of copper on the physical properties of steel is of great importance for the metallurgy of iron.

He further says that when the copper content exceeds about 5 per cent the steel is red short and difficult to forge, if less than 5 per cent is present the steel can either be rolled or forged. He sets forth the advantageous effect of copper on steel and is of the opinion that a further study of copper steel and the determination of the diagram is of importance. He describes in words, and by means of diagrams and tables, the investigation he made.

He states the theoretical consideration of the iron-cementite system, the iron-copper system, and the copper-cementite system and then gives the experimental part of his investigation, using thirteen tables to augment his explanation of the latter phase. He continues by describing the microscopic structures and by means of six photomicrographs shows the result of annealing specimens of varying carbon content and differently etched for a given length of time at 1000 degrees Cent.

The author is of the opinion that the copper added promotes the graphitization of cementite in carbon steel, and says it is interesting to note that the cementite and graphite appear as net-work structures in the photomicrograph. He concludes his paper with a summary of his findings.

GAS PURIFICATION INCREASES OUTPUT OF OPEN-HEARTH FURNACES. By F. W. Sperr, Jr. In *Iron Trade Review*, October 21, 1926, page 1046.

The author stresses the great value of the removal of sulphur from open-hearth fuel gas in the steel industry. He sets forth how the absorption of hydrogen suphide was and is now accomplished, formerly, by dry absorbent material and now by liquid purification. The writer describes one of the latter processes known as the Seaboard process. By means of diagrams the author shows the apparatus used in the liquid purification of open-hearth gas and that used for the recovery of sulphur. The installation of gas purification equipment has made it possible to use coke oven gas through the entire process. This control and elimination of sulphur in the open-hearth process is valuable to the industry. The author is of the opinion that in the future every modern steelworks will have combination ovens heated with producer gas and that the proper place for these gas producers is at the oven and not at the open-hearth furnace. He further says that coke should be the fuel used. This combination, the writer says, will furnish nearly 70 per cent more gas than is now furnished by the ordinary coke oven, or 11,000 cubic feet as against 6600 cubic feet per ton. The purification of this gas can be accomplished economically and checker trouble will tend to be eliminated and the life of the furnace prolonged.

ANNEALING STEEL CASTINGS—I. By Larry J. Barton, in *Foundry*, November 1, 1926, page 854.

The author sets forth the necessity of some kind of heat treatment in specifications for steel castings and is of the opinion that this would result in meeting specified physical tests and relieving strains.

The writer describes from actual experience the various processes used in the making of high grade steel castings to difficult specifications. He is of the opinion that annealing is one of the most important and difficult features. Especially is it difficult to obtain an average anneal throughout when the casting is forty or more feet long or the pieces large, such as are used in railway equipment, locomotive frames, wheels, etc. For annealing, he is of the opinion that the electric annealing furnace is most favorable at the present time. Photographs of the car type annealing oven and the pit type annealing furnace are shown. The pit annealer which consists of a large brick lined pit with a firing chamber on one side and a flue on the other, oil or coal fired, is used for annealing locomotive frames, wheels, etc. While seemingly crude, uniform annealing of large parts can be obtained. The car type annealer is used in many steel foundries.

The author stresses the necessity of temperature control and says that while the optical pyrometer is useful in checking large heats in the pit annealer, it is difficult to obtain true readings as smoke and haze interfere. In scientific annealing these instruments should be considered as aids and not absolute.

He concludes his article with illustrations and a description of the various shaped coupons which are attached to the casting for determining its physical and chemical characteristics and tells where and how these test bars are attached and their purpose for inspection as to tensile strength, uniform fracture and average anneal throughout.

He also shows comparisons of test coupons.

ON THE HARDENING CRACKS IN TOOL STEELS. By S. S. Steinberg. *Messenger of the metal Industry* (Moscow); 5-6 (1926); 70-79.

The author discusses the causes of the formation of cracks during the hardening of tool steel and arrives at the conclusion, that the form of the cementite contained in the steel before hardening has a great influence on the formation of cracks. The more the cementite is divorced in the steel, the more difficultly it passes into a solid solution, the greater the number of centers of crystallization in the shape of grains of cementite remaining in the metal, the less the intensity of hardening and the less the tendency to crack. It is to be concluded therefrom, that the steels containing granular pearlite, especially hypereutectoid with an excess of cementite, are the most proper to avoid cracks.

M. Oknoff and F. Miller, Leningrad, Russia.

GOVERNMENT COOPERATES IN RESEARCH. By Dr. H. W. Gillett. In *Iron Age*, September 9, 1926, page 673.

The author in this paper gives a review of the work of the Bureau of Standards in metallurgical investigations and sets forth how it cooperates with industry and other agencies.

About 60 metallurgical publications, including a wide range of subjects, were made by the Bureau during the past year, of which number about a dozen have been printed in Transactions.

Research projects are chosen with the advice and counsel of Dr. G. K. Burgess, director of the Bureau. A full account of the research associate plan

is given in Bureau of Standards Circular No. 296. This is obtainable from the superintendent of documents. Fee 10 cents.

INDUSTRIAL FURNACES. By Charles Longenecker. In *Iron Trade Review*, October 14, 1926, page 981.

This is one of a series of articles written by the author on the design, construction and function of modern melting, heating and treating units.

The author continues with the subject of melting furnaces. He sets forth and describes the two types of furnaces which are used for melting malleable iron, viz.: air furnace and open hearth. He also shows how the introduction of the use of powdered coal as fuel resulted in the necessity of changing the furnace structure. He is of the opinion that oil is an ideal fuel and gives his reasons.

SURFACE CRACKS ON ROLLED STEEL. By Dr.-Ing. W. Oertel. *Transactions* of the Material Committee of the Verein deutscher Eisenhüttenleute.

This paper deals with the surface cracks on bars and wires of small dimensions from unalloyed or alloyed tool steel, ball steel and high speed steel. The cracks may be:

1. Short cracks,—in most cases only a few hundredths millimeters deep and irregularly distributed on the surface of the material. These are cracks caused by the cooling after rolling; they are very troublesome on all dimensions, being hardened without further treatment, because they grow into hardening cracks at hardening. They may be avoided by putting the pieces in the reheating furnace after the last pass, or cool them slowly in the soaking pits in dry infusorial silica.
2. Long cracks, up to 0.5 millimeters deep, often extending over the whole bar. The author reports the result of an investigation leading to the explanation of these surface defects.

The long cracks are observed on rolled wire from 7 to 5 millimeters thickness and only on pickled material. On raw-rolled and on ground wire there are no cracks; therefore the cracks could only be formed at pickling.

The microscopic examination showed that the rolled wire was really free from cracks and that the material, showing cracks after pickling, contained elongated inclusions of slag, proceeding from the surface. The origin of these cracks may be traced back to the formation of burrs during rolling. By diminution of the pressure of the rolls in the last passes it is possible to limit the occurrence of the pickling cracks to a minimum.

Abstracted by Dr. Hans Pollack, Germany.

ON THE A_2 LINE IN THE EQUILIBRIUM DIAGRAM OF THE IRON-CARBON SYSTEM. By Kotaro Honda. Reprinted from the *Science Reports* of the Tohoku Imperial University, Series I, Vol. XV, No. 2. Sendai, Japan. May, 1926.

The author in this article briefly describes the A_1 and the A_2 lines and points out the distinction between them by means of a diagram. He also gives consideration to the question—"What are alpha and gamma irons?

ON THE CHANGE OF THE HARDNESS AND THE DENSITY IN IRON AND STEEL CAUSED BY COLD-WORKING. By Toyozo Ishigaki; *Kinzoku no Kenkyu,* Vol. 3, No. 7, 1926, page 337-349.

The author made the following investigations:—

(a) the relation between the increase of hardness and the degree of cold working;

(b) the relation between the change of density and the degree of cold working;

(c) the effect of tempering at different temperatures on the hardness of cold-worked specimens.

(d) the effect of the tensile test at high temperatures on the hardness of steel at room temperature.

The specimens used were Armco iron and Swedish steels containing different amounts of carbon; the cold working was given by a tensile test machine, a hammering and a compressing up to 235,000 pounds per square inch. The results of experiments are summarized as follows:—

(1) By cold working, the increase of hardness is greatest in Armco iron and diminishes in steel with increasing content of carbon.

(2) For a given steel, the hardness increases with the degree of cold working, tending to an asymptotic value.

The above change of hardness is due to the internal strain caused by cold working.

(3) By tempering a cold-worked specimen at a gradually increasing temperature, the hardness slightly decreases to about 100 degrees Cent., then begins to increase and attains a maximum at 270-330 degrees Cent. It then decreases, but from 400 degrees Cent., this diminution becomes somewhat slow, and in approaching 600 degrees Cent., the hardness rapidly decreases to attain the normal value before cold working.

The increase of hardness by tempering in the vicinity of 200-300 degrees Cent. is due to the stabilizing of a more unstable portion during cold working by virture of thermal agitation. By this stabilizing the slipping is made more difficult and consequently the hardness is increased.

The slow decrease of hardness between 450-550 degrees Cent. is due to the hardening by a crystal refining accompanying recrystallization, and its subsequent rapid fall is caused by the release of inter-strain.

(4) The effect of the tensile test at high temperatures on the hardness of steel at room temperature is very similar to the effect of tempering on the hardness of cold-worked specimens and can be explained in the same way as before, the only difference being that in the former case, a part of unstability is removed during tempering, while in this case, its removal takes place during the stretching at high temperatures.

(5) In the case of a specimen broken by a tensile test, the density of the specimen decreases, the amount of which increases with the degree of cold working. In the case of compression, the density of a speci-

men decreases at first, reaches a minimum and then begins to increase; after passing through a maximum, the value of which is less than the initial value, it begins to decrease.

The diminution of density is a combined effect of two causes:—the first cause is minute cavities occurring in slip planes or crystal boundaries as proposed by Prof. G. Tammann, and the second to the interatomic expansion caused by the interstrain as proposed by Prof. K. Honda[1]. The above change of density by cold working is satisfactorily explained as the combined effect of these two.

(6) By annealing cold-worked specimens, their density remains nearly constant up to 300 degrees Cent., and then begins to decrease up to 550 degrees Cent., after which it again increases towards its initial value before cold working. The decrease of density is probably due to the crystal refining due to recrystallization, by which the total surface of microcrystal, and consequently minute gaps in the boundaries, is increased. Subsequent increase of density is due to a gradual vanishing of these minutes gaps of cavities and the release of the interstrain.

Abstracted by Dr. Kotaro Honda, Japan.

ON THE ELECTRICAL RESISTANCE OF MOLTEN METALS AND ALLOYS. By Yoshiji Matsuyama. *Kinzoku no Kenkyu*, Vol. 3, No. 9, page 439-455.

The object of the present paper is to measure the electrical resistance of binary systems in their liquid state, and from the resistance-concentration curves, to decide the question, whether an intermetallic compound exists as such in its molten state, or whether it exists in atomic state being completely dissociated. The method of experiment was the same as that given in the first report (*Kin. Ken.* Vol. 3, No. 4, page 254), and the results of observation may be summarized as follows:—

(1) In binary alloys, the components of which form no compounds with each other in solid state, the resistance-concentration relations can be expressed by a curve bending slightly from one component to another without showing any maximum or minimum.

(2) In binary alloys, the components of which form one or more intermetallic compounds with each other, the resistance-concentration curve shows a maximum in the concentration of the compound; and the temperature coefficient-concentration curve shows a minimum in the same concentration.

(3) Usually the maximum and minimum are very sharp; but in some cases it is more or less rounded, showing some degree of dissociation of the compound.

(4) From the above results, it is to be concluded that in the iron-carbon system, the cementite (Fe_3C) is present as such in the molten alloys, though some small degree of dissociation might result.

Abstracted by Dr. Kotaro Honda, Japan.

[1] *Journal*, Institute of Metals, Vol. 32, No. 2, 1924, page 412.

ON BALL-BEARING STEEL. By Dr. E. Houdremont and H. Kallen, Krefelder Stahlwerk A.-G. Krefeld. *Zeitschrift des Vereines Deutscher Ingenieure,* July 31, 1926, page 1035-1039.

In this paper the authors discuss the manufacture, the structure and the defects of ball bearing steel, containing about 1 per cent carbon and 1.5 per cent chromium.

Ball bearing steel is molted in the open-hearth furnace or in the electric melting furnace. The crucible steel process is not discussed on account of the high costs of making it that way.

The charge for melting this steel may be either ordinary pig iron and scrap, containing a large amount of impurities and dangerous components as phosphorus and sulphur, which require a careful refining process, or the charge may consist of Swedish charcoal pig iron or pure scrap, free from impurities. In this case the manufacture may be confined to a melting down and alloying.

The chromium increases the hardness and the elastic limit of the steel. These improvements in conjunction with the grain-refining effect of chromium on the structure of steels is to increase the wear-resistance of ball bearing steel over that of a carbon steel. The presence of chromium, causes the carbides to be more finely distributed in the matrix than in carbon steel. Chromium does not effect only the size and the distribution of the carbides but increases also their number. In the hardened matrix of a ball bearing steel there is a great number of free carbides embedded. Chromium increases the hardness depth of a steel. Carbon steels with 1 per cent carbon and a cross section greater than 10 x 10 millimeters show after quenching from the normal hardening temperature a hardened case and a tough core. With the addition of chromium this core disappears more and more and at an amount of 1.5 per cent chromium its influence makes itself intensively conspicuous. For this reason chromium ball bearing steel may be hardened in oil instead of in water and by oil quenching hardness defects such as warping and cracking may be avoided.

Of great importance is the rolling or drawing temperature. Too long and too high heating is detrimental to the properties of the steel. There are two sorts of incorrect treatment: the burning and the overheating of the steel structure. Burnt steel may never be reclaimed, while overheating of the structure may be thoroughly removed by suitable heat treatment.

Improper choice of the hardening temperature may impair the duration of life of a bearing to a high degree. Insufficient hardening causes an imprinting of the balls and too high hardening causes brittleness, deficient elasticity, coarse-grained fracture, etc.

The best properties of the steel, great hardness and best toughness, may be obtained by a double hardening. The temperature of the first hardening is about 50 degrees Cent. (90 degrees Fahr.) higher than that of the following—one, taking place at normal hardening temperature.

Recently ball bearing rings have been made from forged rings as well as drawn tubes; a difference in the quality between both of these materials was not studied.

Abstracted by Dr. Hans Pollack, Germany.

ON THE EQUILIBRIUM BETWEEN AUSTENITE AND CARBON OXIDES. By Genske Takahasi. Reprinted from the *Science Reports* of the Tohoku Imperial University, Series I, Vol. XV, No. 2. Sendai, Japan. May, 1926.

The writer of this article describes and illustrates the apparatus used in his investigations and the manner in which the specimens were prepared. He then describes the method of the experiment making it concrete by the use of graphs. He tabulates the results of experiments giving the temperature of heating, composition of gas, carbon in specimen before heating, duration of heating, velocity of gas, carbon in specimen after heating and comments. In his discussion of the results of his experiments he sets forth his findings, especially those regarding carbon content, with rising temperatures. The author also adds to the interest of the paper by incorporating 23 photomicrographs.

ON THE MODULI OF ELASTICITY AND RIGIDITY, AND THEIR CHANGE CAUSED BY MAGNETIZATION, IN DIFFERENT KINDS OF STEEL. By Kotaro Honda and Tomoichi Tanaka. Reprinted from the *Science Reports* of the Tohoku Imperial University, Series I, Vol. XV, No. 1. Sendai, Japan, April, 1926.

The authors set forth the results of their investigation of the measurement of the moduli of elasticity and rigidity for different kinds of steel, such as carbon, nickel and cobalt steels and their change by magnetization. They describe and illustrate the apparatus of measurement used for this investigation and the manner in which the measurements were made. The authors measured the moduli of elasticity and rigidity of carbon steels in the annealed and quenched states. A summary of their findings is set forth.

ON THE SPECIFIC HEAT OF CARBON STEELS. By Saburo Umino. Reprinted from the *Science Reports* of the Tohoku Imperial University, Series I, Vol. XV, No. 3. Sendai, Japan, July, 1926.

The author in his introduction sets forth the investigations which have been made by various investigators along this line. He then describes the arrangement and method of experiment used by him in his experiments; the preparation of the specimens; and gives detailed results of the experiments. In a short summary he states the conclusions at which he arrived.

ON THE MALLEABLE CAST-IRON AND THE MECHANISM OF ITS GRAPHITIZATION. By Tario Kikuta. Reprinted from the *Science Reports* of the Tohoku Imperial University, Series I, Vol. XV, No. 2. Sendai, Japan. May, 1926.

The author in this paper describes his investigations dealing with the graphitization of white iron into malleable cast iron; the effect of temperature on the first and second stages of graphitization; the effect of thickness of castings on the first and second stages of graphitization; the effect of tapping temperature on the graphitization; the effect of various elements on the graphitization of malleable cast iron (silicon, carbon, manganese, sulphur and phosphorus). He gives some practical tests especially along the lines of heat treatment and sets forth his findings in tables. He then takes up the

subject of the growth of malleable cast and shows the results of repeated heating and cooling. He concludes his paper with a consideration of the mechanism of graphitization of white cast-iron. His summary and numerous graphs are of assistance to the reader.

ON THE HEATS OF TRANSFORMATION AND FUSION IN CARBON STEELS, CAST IRON AND SOME OTHER METALS. By Saburô Umino. *Kinzoku no Kenkyu,* Vol. 3, No. 8, pages 385-393.

By means of the method of mixtures, the author measured the heat of transformation and that of fusion for carbon steels, cast iron, manganese, and thallium. Steels contain 0.57, 0.94, 1.10 per cent carbon; and cast iron 4.31 per cent carbon and 1.11 per cent Silicon.

(1) Heat of A_0 transformation in steels (per gram):

Per Cent Carbon		
0.57	215°	0.86 gr. cal.
0.94	215°	1.38 gr. cal.
1.11	215°	1.63 gr. cal.
		1.48 cal. for 1% C.

(2) Heat of fusion of the cast iron: 46.43 gr. cal.

(3) Heats of transformations of manganese:

835°	2.88 gr. cal.
1044°	4.53 gr. cal.
Latent heat of fusion:	64.80 gr. cal.

newly discovered allotropies of manganese.

(4) Heat of transformation and of fusion in thallium:

232°	0.60 gr. cal.
303° (melting point)	3.67 gr. cal.

Abstracted by Dr. Kotaro Honda, Japan.

ON THE EQUILIBRIUM DIAGRAM OF THE SYSTEM FeS-MnS. By Zen'ichi Shibata. *Kinzoku no Kenkyu,* Vol. 3. No. 8, page 403-408.

By means of thermal analysis and microscopic investigation, the author determined the equilibrium diagram of the system FeS-MnS as shown in Fig. 1. He found the following facts:

(1) The melting point of FeS is 1163° ± 2°C; and that of MnS 1610° ± 3°C.

(2) The author could not confirm the existence of a compound $Fe_3Mn_2S_5$ found by Röhl.

(3) FeS and MnS form limited solid solutions alpha and beta with each other.

(4) Alpha solid solution dissolves 2 per cent MnS at the eutectic point, but almost none at room temperature. On the other hand, beta solid solution dissolves 75 per cent of FeS at the eutectic point, but only 24 per cent at room temperature.

Abstracted by Dr. Kotaro Honda, Japan.

News of the Chapters

STANDING OF THE CHAPTERS

IN the following tabulation there appears the relative membership standing of the 30 chapters of the Society as of November 1, 1926. On the next page appears the relative standing of the chapters as of September 1, 1925, and September 1, 1926, compared with the standings as of June 1, 1925, and June 1, 1926.

STANDINGS OF THE CHAPTERS AS OF NOVEMBER 1, 1926

GROUP I	GROUP II	GROUP III
1. Cleveland 401	1. Lehigh Valley ..138	1. Los Angeles 71
2. Chicago 376	2. Hartford 133	2. New Haven 69
3. Philadelphia 358	3. Milwaukee 122	3. Washington 68
4. Detroit 358	4. Golden Gate 119	4. Tri City 64
5. Pittsburgh 316	5. Syracuse 95	5. Rockford 63
6. New York 265	6. Cincinnati 95	8. Rochester 59
7. Boston 244	7. Indianapolis 84	7. Montreal 58
	8. St. Louis 82	8. Worcester 58
	9. Buffalo 60	9. Toronto 56
	10. Northwest 55	10. Schenectady 49
		11. Providence 44
		12. Fort Wayne 40
		13. Springfield 30

There were 153 new and reinstated members during October. Thirty-two were dropped for non-payment of dues, and eleven on account of resignations and deaths, leaving a net gain for the month of 110 members. The total membership of the Society at the present time is 4217.

GROUP I—Cleveland gained 8 members last month, which with Chicago's substantial increase of 17, makes a difference of but 25 members between the two chapters. All of the other chapters in this group added a few to their roster; Philadelphia and Detroit being tied with 358 members each.

GROUP II—Though Lehigh Valley lost 4 members last month, they still hold position No. 1 in this group. Hartford, Golden Gate, Cincinnati, St. Louis and North West all showed gains; the others either hold the same place or give evidence of a loss.

GROUP III—With the exception of Los Angeles and Schenectady, all of the other members of Group III showed gains. Rochester and Montreal, by gaining three new members each, have shot ahead of Worcester, which occupied 6th place last month, and they now hold positions 5 and 6. Worcester even with an increase drops back to No. 8. The other chapters retain the same relative positions as on the previous report.

CHAPTER STANDINGS ON THE FOLLOWING DATES

	June 1 1925	Sept. 1 1925	Loss or Gain	June 1 1926	Sept. 1 1926	Loss or Gain
Boston	223	215	— 8	234	238	+ 4
Buffalo	50	52	+ 2	58	59	+ 1
Chicago	265	263	— 2	312	335	+23
Cincinnati	76	74	— 2	83	87	+ 4
Cleveland	314	324	+10	376	277	+ 1
Detroit	360	376	+16	386	349	—37
Golden Gate	93	88	— 5	129	122	— 7
Hartford	126	121	— 5	138	133	— 5
Indianapolis	55	53	— 2	82	85	+ 3
Lehigh Valley	98	101	+ 3	137	141	+ 4
Los Angeles	61	60	— 1	74	74	..
Milwaukee	70	71	+ 1	100	104	+ 4
New Haven	56	56	..	56	57	+ 1
New York	203	209	+ 6	255	254	— 1
Northwest	44	43	— 1	48	46	— 2
Philadelphia	260	256	— 4	356	359	+ 3
Pittsburgh	292	295	+ 3	317	305	—12
Rhode Island	39	39	..	43	40	— 3
Rochester	62	59	— 3	55	53	— 2
Rockford	41	42	+ 1	62	60	— 2
Schenectady	51	51	..	47	50	+ 3
Springfield	31	30	— 1	30	29	— 1
St. Louis	60	59	— 1	76	75	— 1
Syracuse	71	74	+ 3	101	96	— 5
Toronto	37	39	+ 2	46	47	+ 1
Tri City	66	65	— 1	66	62	— 4
Washington	50	48	— 2	61	62	+ 1
Worcester	47	47	..	57	56	— 1

The Society as a whole shows a very consistent and steady growth since its inception. This does not hold true for all our chapters. A review of the chapters' standings shows a few rather marked ups and downs. The cause of these rises and falls is best known to the chapters themselves.

June 1, usually marks a recession in chapter activities. About September 1, this work is again resumed. This is, of course, as it should be.

A comparison of membership standings for June 1, and September 1, 1925, also June 1, and September 1, 1926, is shown on the upper part of this page. The loss or gain is shown in each case.

In the summer, this "turnover" is higher than at any other season. For some chapters, are not these figures rather high?

Business men seek to reduce their labor turnover to a minimum. Does not the same principle apply to the members of the American Society for Steel Treating? Let's follow their lead.

Would it not be well for the membership committee of chapters sustaining heavy losses to analyze the cause of these losses. A suggestion is offered that a strong appeal be made to these former members seeking to bring them back into the fold.

BOSTON CHAPTER

The October meeting of the Boston Chapter was held at the Fitchburg,

Mass., plant of the Simonds Saw and Steel Company, Friday, October 1, 1926. The afternoon was spent in the plant visiting the departments manufacturing hand saws, circular wood and metal saws, band saws, high speed steel knives of many kinds and similar products. The hardening room was completely equipped with many types of special furnaces, all under pyrometer control, and the greater part of the afternoon was spent in this department and the forging plant, which adjoins it.

One of the most interesting operations was the heat treatment of large circular saws, many of them running up to six feet in diameter. The large circular plates were heated in oil-fired furnaces and with the aid of special handling tools, were quenched on edge in oil with little effort. The care in heating and quenching, together with an initial steel of fine quality resulted in a saw with very little distortion, considering the magnitude of the operation. The Simonds method of tempering under pressure between forming dies enables them to temper and flatten simultaneously, thus eliminating straightening with a hammer, which they claim is detrimental to the efficiency of the saw. The finished saw is subjected to a hand hammering operation known as "tensioning", it being tensioned, or dished, throughout the portion between the mandrel hole and the rim, to counterbalance the pull caused by the centrifugal force when the saw is speeded up in the mill. The larger the plate and the higher the speed, the greater the required tensioning. Some of the other interesting jobs were the quenching and tempering of hand saws, the welding of high speed steel to carbon steel in the manufacture of large knives and the hardening of the finished knives.

Exactly one hundred members and their friends were guests of the Simonds Saw and Steel Company at a chicken dinner, served at the Hotel Raymond at 6:00 p. m. Immediately following the dinner, the secretary gave a report of the nominating committee meeting at Chicago and A. H. D'Arcambal, of the Hatrford Chapter, was requested to say a few words regarding the Chicago Convention from an exhibitor's point of view.

A. M. Remington, consulting engineer for the Simonds Company, under whose direction the visitation was held, introduced the several executives of his Company and then presented the president, Alvin T. Simonds, who gave a very interesting talk on his pet subject "Economics." Mr. Simonds handled his subject as much as possible from the standpoint of the heat treater and metallurgist and urged those present to make some study of economics to aid them in their business and home life. Following his talk, Mr. Simonds was asked to answer many questions and spoke at some length on the purchase of metallurgical materials.

After a few words by Dr. Waterhouse, regarding the latest heat treating literature, the members adjourned at 9:00 p. m. *H. E. Handy.*

CINCINNATI CHAPTER

The Cincinnati Chapter of the American Society for Steel Treating held its second regular meeting at McMicken Hall, University of Cincinnati, on the evening of October 6, 1926. Dr. William M. Guertler delivered an address on "The Systematic Procedure for Determining the Limits of Utilization of New Metal Combinations in Technical Practice." Dr. Guertler presented

the results of many years of research on binary and ternary alloy systems. These studies have resulted in important developments in high strength alloys as well as in corrosion resistant metals and have led to valuable generalizations as to the behavior and properties of complicated alloy systems.

Dr. Guertler covered a very extensive field in his talk and illustrated his points with a large number of slides. So many ideas were presented that some of the treaters suffered from mental indigestion.

The local chapters of the American Society of Mechanical Engineers and the American Chemical Society co-operated with the American Society for Steel Treating in making this a very successful meeting. The attendance was 247.

The Cincinnati Engineers Club will hold a joint meeting with the American Society for Steel Treating on November 18, in the new Engineers Club Building. Mr. E. C. Bain, research metallurgist of the Union Carbide and Carbon Research Laboratories, will speak on some of the new developments in high chromium alloys, as well as some of his recent X-ray work.

CLEVELAND CHAPTER

As the world's largest producer of electric furnace steel, The Timken Roller Bearing Company proved a magnet of sufficient strength to draw ninety-six members of the American Society for Steel Treating to the Timken plant Friday afternoon, October 22. The visitors were shown through the steel plant in which not only alloy steel for Timken bearings is made, but also a high grade product for outside consumers.

The visitors were impressed particularly with the magnitude of Timken facilities. A briquetting machine which compresses the turnings and borings into a solid mass or briquette was the first machine of its kind ever built. By the use of this machine the small steel shavings from the various machines are reclaimed. This machine was built for use in the Timken plant and cost $125,000.00. This amount was saved during the first year's operation.

Five 6-ton Heroult electric furnaces are used to melt down the scrap steel and the various alloys. The capacity of these furnaces is sufficient to supply the steel necessary for the manufacture of Timken bearings as well as the tonnage sold to outside consumers. The 35-inch blooming mill, modern in all details, was of more than ordinary interest to the visiting party. It is in this mill that the ingots brought from the electric furnaces are rolled into blooms for the 22-inch merchant mill. The bridge truck wheels, trolleys, shafts and sheaves of the 10 and 15-ton Morgan cranes in the 35-inch mill are equipped with Timken bearings.

The remaining units, such as the reheating furnaces, the 16-inch, 12-inch and 10-inch rod and bar mills, the piercing mill and cold drawing departments were visited. Following the trip through the plant, the party adjourned to Brookside Country Club where dinner was served and the regular monthly meeting of the society was held.

After the dinner Hugh Rodman, president of the Rodman Chemical Company of Verona, Pa. delivered an address on, "Case Hardening Materials."

Chairman W. P. Sykes presided.

Mr. Rodman divided his subject into two sections i. e., the various types or kinds of solid case hardening compounds and the making of these compounds.

The address was very instructive as he explained the action of the various energizers used in case hardening compounds. Also the discussion afterwards brought in some enlightening facts.

After the meeting Mr. Sanford of The Timken Roller Bearing Company said a few words. Dr. Albert E. White of University of Michigan, who happened in Canton on business also said a few words.

A. E. Buelow.

DETROIT CHAPTER

The activities of the Detroit Chapter are off to a fine start. The first meeting of the year was held Wednesday, September 29, in the General Motors Building.

It has been our policy, where possible, to have some internationally known man talk here each year and this time it was Dr. William M. Guertler, who talked on light metal alloys. We feel that with the announced policy of the society to expand its interests into the field of none-ferrous metallurgy that it was better indicated with his talk than in any other way. More significant than the inclusion of this subject was the hearty discussion that followed.

The Detroit Chapter has had of late, in addition to the regular technical paper, some one to talk for a few minutes just after our dinner on something of general interest. We were fortunate this time in having for the "Coffee Talk" Captain Eddie Rickenbacker. The captain was well known as the Ace of Aces in the American army and talked to us of the future of aeronautics. To be sure we in Detroit are well awake to the glory of such dreams as Captain Rickenbacker's fancy painted to us of ships and planes of the future but not all of us dare to voice such ideas in the highly colorful way that he did.

Dr. Guertler's lecture was the same that he gave in Chicago and we expect that it will be printed in full in the TRANSACTIONS. *J. L. McCloud.*

GOLDEN GATE CHAPTER

This will report the September meeting of the Golden Gate chapter of the American Society for Steel Treating. There was a total of fifty-three present at the meeting which was held on September 15, at the Engineers' Club, San Francisco, President Drake presiding. The meeting was preceded by the usual dinner.

Mr. Garlinger announced the intention of distributing blue prints announcing each monthly meeting, to be posted in industrial plants, in order to attract attention and thereby increase the turnout at these meetings. It was announced by chairman Drake that our representative to the Chicago Convention, Mr. Gearhart, had been appointed to open the first technical meeting at the convention.

The first paper was presented by S. H. Edwards, of the Industrial Service Company, who made a review of the articles in TRANSACTIONS for April and May, entitled "Facts and Principles Concerning Heat Treatment of Steel; Parts 5 and 6". These articles were written by H. B.

Knowlton, member of the Milwaukee chapter, and an instructor in metallurgy at the Milwaukee Vocational School, Milwaukee.

The articles which appeared as Part 5 in TRANSACTIONS for April dealt with steel failures. This article was ably discussed by Mr. Edwards. The article covered in a brief way the mechanism of steel failures, and took up the application of theory to practical problems in the interest of preventing cracking, warping, and breaking of steel parts. As a means of bringing out these applications, it was necessary to discuss the composition of matter, and the properties and construction of crystals, which exist in various conditions in steel and iron.

Part 6 of this discussion appeared in the May issue of TRANSACTIONS and took up distortion, warping and cracking of steel during heat treatment.

The second talk of the evening was devoted to "Discussion of the Dilatometer" by C. S. Moody, metallurgist of the Caterpillar Tractor Company. This paper was accompanied by slides projected from the spectroscope. The particular dilatometer which was discussed and which was illustrated, is that manufactured by the Stanley P. Rockwell Company. This instrument magnifies the dilation of steel by about 100 times.

Mr. Moody explained that the world "dilatometer" means measuring of expansion. When applied to heat treating, it signifies the measuring accurately of elongation and contraction of steel under varying temperatures. By magnifying the change in shape of the piece of steel under temperature change, it is a simple matter to record the variations on a chart, which would indicate the exact time in which the piece of steel undergoes the critical temperatures. Mr. Moody pointed out that in a number of cases which were tried out, that this method assisted materially in determining quenching temperatures. *C. R. Owens.*

HARTFORD CHAPTER

The regular monthly meeting of the Hartford Chapter was held Tuesday evening, October 12, in the Assembly Hall of the Hartford Electric Light Company. The speaker of the evening was G. A. Richardson, manager of the technical publicity department of the Bethlehem Steel Company, Bethlehem, Pa. The speaker chose for his subject, "Manufacturing High Quality Steels" and presented an interesting lecture. The lecture was illustrated throughout with motion pictures, first showing the operation of blast furnaces which poured heats of 100 to 125 tons of molten metal every five hours. To make one ton of pig iron it required eight tons of material of which four tons were air.

The Bessemer and open-hearth methods were then shown, the open-hearth having a capacity of 200-250 tons of metal. It could be seen that one of the successes of making steel was the use of automatic machinery. The blooming mill was well illustrated and the speaker explained its operation thoroughly.

The crucible process and electric furnace method of manufacturing steel were then taken up. The speaker gave several statistics as to the amount of steel made by the electric furnace. 1908 produced 55 tons while

in 1918, ten years later, over half a million tons were made. In showing the later films it could be seen how much more labor is involved in the making of tool steels as compared with steel made from the Bessemer or open-hearth furnace.

The last film to be shown was one on the manufacture of car wheels and large ring forgings, and the amount of stock that has to be carried by a manufacturer to successfully take care of this line. The lecture, elementary to some of us, was, nevertheless, made very interesting, and as it was put on as an educational one it proved very instructive to all.

The Hartford Chapter held a special meeting on Wednesday evening, October 20. Dr. William M. Guertler was the speaker of the evening. Preceding the meeting Dr. R. Woodward, chairman of the chapter, introduced F. G. Hughes, vice president of the New Departure Manufacturing Company, who is nominee for vice president of the national society. Mr. Hughes spoke on the position that the American Society for Steel Treating holds among the other national technical societies. He also spoke very forcibly on the outstanding aims of our society and that the greatest aim was to unite the research man with the man in the shop, which no other society has accomplished as successfully as the American Society for Steel Treating. He then spoke about the next convention and exposition of the society to be held in Detroit and gave reasons why we should support and aim to make this show a success. His closing remarks were that the member receives three times more from the society than he pays as his dues. Mr. Hughes then introduced the speaker of the evening, Dr. Guertler, who gave his lecture on "The Corrosion Resistance of Alloys" and divided his talk into five parts.

1—Solid Solution
2—Electrolitic Action
3—The Avoidance of Catalytic Agents
4—Protective Coatings
5—Noble metal, the most protective resistance

Dr. Guertler explained each of the above subjects very thoroughly and in such a manner as to make it very interesting. The members were very much surprised that the doctor could deliver a lecture of this type in such an interesting manner. A lively discussion was taken part in after the lecture. The questions were answered by Dr. Guertler in a very satisfactory manner, especially the reason for the stainless properties of stainless steel after hardening and not before. In the discussion it could be seen that the doctor was able to discuss very thoroughly any of the metallurgical problems put before him. The chapter was very proud to have the doctor appear before it and give such an interesting talk and invitations were sent to all the local engineering societies. Also the National Society should be congratulated for obtaining lectures of Dr. Guertler's caliber.

H. J. Fishbeck.

LOS ANGELES CHAPTER

The last regular meeting of the Los Angeles chapter was held September 16, at 2:00 p. m. The program committee had arranged, through the

courtesy of the Southern California Iron and Steel Co., for an inspection trip through its plant. H. W. Schmid, assistant chemist, was delegated to escort the members and visitors, thirty-six in all, through the various departments.

We first inspected the open-hearth department, which consists of three, 35-ton basic open-hearth furnaces. Next we passed from the open-hearth floor down into the pit and out to the ingot yard.

Our attention was next called to the soaking pits and forge press which have a capacity of 1000 tons. This press is used mainly for forging rounds, flats and squares, prior to rolling.

We next passed by the billet or bar mill where four inch squares were being rolled for use in the finishing or twelve inch mill. The electrical equipment for this twelve inch mill is of the latest type.

Rounds, flats, squares and reenforcing bars are rolled in this mill.

Our visit in the machine shop was very interesting. This shop is maintained purely as a maintenance shop for dressing and turning rolls and taking care of repairs to equipment.

The last department and one of the most interesting the members visited was the bolt department where bolts, nuts, rivets and lagg screws are manufactured.

The tour of the plant ended in this department, and all present were of one accord, that this meeting held at the plant while in full operation, was one of the most interesting held for some time.

After a vote of thanks to Mr. Schmid and the officials of the company the meeting adjourned at 4:30 p. m. *E. C. Black.*

MILWAUKEE CHAPTER

Milwaukee Chapter's second meeting of the season was held October 22 at the Blatz Hotel.

The 6:30 dinner, as well as the lecture, was well attended. Jordan Korp, of the Leeds and Northrup Company, was the speaker of the evening.

Mr. Korp's talk on heat treating and hardening of tools was very instructive as well as interesting, as was evidenced by the amount of discussion.

M. E. Greenhow.

NEW HAVEN CHAPTER

The New Haven Chapter held its regular monthly meeting at the General Electric Company, on Thursday, October 14, at Bridgeport, Connecticut. It was surely a big afternoon and evening.

The festivities started at 3 p. m. with "traffic officer" H. D. Smith of the Colonial Steel Company doing traffic duty at the corner of Boston Avenue and Bond Street.

Eighty-five members and friends enjoyed the plant visitation from 3 to 5 p. m. The only trouble the writer found with this part of the day was that the 5 o'clock whistle blew too soon.

At a few minutes past five, about forty journeyed over to the Bridgeport Brass Company to witness the pouring of a heat of brass. The remainder of the crowd were entertained until half past six by Kelly and

Ryan, the melody boys. From the amount of applause, this part of the program was heartily enjoyed, thanks to Duley.

At half past six, escorted by H. G. Wells, we journeyed to the cafeteria of the General Electric Company where seventy-two men did justice to a nice steak dinner.

We returned to the Assembly Room shortly before eight and promptly on the hour, the meeting was called to order by our chairman, C. J. Sauer. Several of our new members were introduced and received the glad hand from those present. Fred Dawless, chairman of the membership campaign gave a short talk telling of the progress made so far and winding up with a very good golf story. The writer who served on the National Nominating Committee was called on for a few remarks concerning the Chicago Exposition. It is my opinion that this was an attempt to find out where I disappeared to in Chicago during the convention. Well I am like all other steel treaters and will say, "Try and find out."

Following the short business meeting, C. L. Ipsen of the General Electric Company, Schenectady, New York, gave a very interesting talk on "Electric Furnaces for Steel Treating." This talk was illustrated with lantern slides and from the amount of discussion following there is no doubt but that the talk was well enjoyed. This talk was attended by one hundred and twenty-five men which goes to show that this meeting was the best ever held by the New Haven Chapter. The success of this meeting was entirely due to the committee in charge, viz: Raymond T. Porter, Heppenstall Forge Company; H. G. Wells, General Electric Company; and Paul Clark, Bridgeport Brass Company. Many thanks to these men for the capable manner in which they handled the details of this successful meeting.

The attendance prize donated by Russell S. Young of the O. K. Tool Company, Shelton, Connecticut, was won by Forest G. Purinton of the Patent Button Company, Waterbury, Connecticut.

The R. Wallace & Sons Manufacturing Company, Wallingford; the New Haven Clock Company, New Haven, and the Heppenstall Forge Company, Bridgeport were all present in good numbers. The old guard can always be depended on for their usual good turn out.

The meeting closed shortly after ten and we were soon on our way home. Tired but happy. *Walter G. Aurand.*

NORTHWEST CHAPTER

The second meeting of the Northwest Chapter for this year was held on Wednesday evening, October 20. Professor G. C. Priester gave a paper on the Effect of Annealing on the Machinability and Physical Properties of Cast Iron. In this paper Professor Priester gave some interesting results that he obtained while studying the properties of cast iron. The annealing increased the ease of machining in all cases while the tensile and transverse strengths were decreased. The deflection in the transverse test was not affected to any appreciable extent by the annealing.

A study of the stress—strain diagrams showed that cast iron had a

definite elastic limit. This property of cast iron had not been previously reported in the literature.

Photomicrographs were shown to illustrate the change in structure which took place during the annealing process. They showed that the cementite decomposed to form graphite and iron. The paper was well illustrated by means of slides and the results interpreted in such a way that they were of use to the practical man.

Prof. O. W. Potter gave a comprehensive report on the Foundrymen's Convention at Detroit. He gave short abstracts of the papers that were presented and discussed the great interest that the metal industries are taking in the study of these problems. Professor Potter emphasized especially the goodwill that was prevalent at the convention between foundrymen, both from this country and foreign countries.

Albert Zima gave a report on the good fellowship at the convention in Chicago. According to his report, the convention succeeded very well in promoting good fellowship amongst the various chapters.

Dr. O. E. Harder announced that Dr. Guertler would be in Minneapolis November 8 and 9, to deliver three lectures. It was possible by cooperation of the steel treaters Society with several other technical organizations in the Twin Cities to bring this noted metallurgist to Minneapolis. A large turn-out is expected to hear Dr. Guertler's lectures.

O. P. Briggs, our latest sustaining member, was called upon for a few remarks. He congratulated the chapter on the good work it is doing and emphasized the importance of good will and cooperation.

After the meeting, old acquaintances were renewed by those who had been absent during the summer. The great interest and spirit shown at this meeting indicates that the Northwest Chapter should have a very successful year.

L. J. Weber.

PHILADELPHIA CHAPTER

The first meeting for the year of 1926-27 was held by the Philadelphia chapter on Thursday evening, October 7, at the Engineers' Club, 1317 Spruce Street. The meeting was preceded by a dinner at the same place, which was attended by thirty-five members and guests.

The meeting was called to order by the chairman, W. B. Coleman, at 8:15 o'clock and virtually opened with a "Bang." Over two hundred members and friends of the chapter were present and it was necessary to find extra chairs so that all present could be seated.

The business session of the meeting brought out some interesting features concerning the course being given at Temple University under the auspices of the chapter, and which now enters upon its sixth year. The chairman named the personnel of the various committees and stressed the establishment of the new committee which has been named the Committee for the Retention of Members.

The speaker of the evening was F. B. Lounsberry, vice-president of the Atlas Steel Coropration of Dunkirk, N. Y. The subject chosen by Mr. Lounsberry was "Tool Steel Failures, Their Causes and Cures." This talk was profusely illustrated with splendid lantern slides, each of which speci-

fically illustrated the points that were stressed. Mr. Lounsberry mentioned among many things, that of all complaints sent into his plant concerning the steels sent to different customers, fifty-five per cent were traced to manufacturing, thirty-six per cent to improper handling by the customer and the balance to general causes. This caused some comment, coming as it did from the head of such a large tool steel plant. He also discussed the use of the macro etch, the annealing, the machining, and treating of tool steels. His dissertation on high speed steel left nothing to the imagination or in the realm of doubt.

The discussion which followed this paper showed the keen interest that was taken in its presentation, and the Philadelphia chapter unhesitatingly recommends that as many chapters as possible should try to have Mr. Lounsberry on their programs during the next few months.

Friday, October 8, a number of the members of the chapter journeyed to the Aberdeen Proving Grounds at Aberdeen, Maryland, as guests of the Army Ordnance Association and enjoyed a most instructive as well as entertaining day watching the firing of the large guns, and the aeroplane manoeuvres, etc. The thanks of the Society are certainly due the Army Ordnance Association for having extended the invitation to participate in their big event. *Arthur W. F. Green.*

PITTSBURGH CHAPTER

In as much as Dr. William M. Guertler was scheduled to give three lectures at the Carnegie Institute of Technology on October 11, 12 and 13, the executive committee decided that, as the members would want to attend these lectures, there would be no other meeting held during October except that the chapter officers and members should assemble and have dinner with Dr. Guertler in the dining hall of the United States Bureau of Mines on the evening of the 12 at 6:30. This was done and, after dinner, those assembled accompanied Dr. Guertler to the Carnegie Union where his lecture was given. *Harry A. Neeb, Jr.*

ROCHESTER CHAPTER

On September 27, the Rochester Chapter of the American Society for Steel Treating held its first meeting of the year at the Mechanics Institute. Burton DeLong of the Carpenter Steel Company gave a very interesting and instructive talk on high speed steel. Prior to the talk the monthly dinner was served to the members.

October 11, the Rochester Chapter was given a talk on high speed steel by J. P. Gill of the Vanadium Alloys Company.

William M. Guertler presented one of his seven papers to the Rochester Chapter on October 14, at the Eastman Kodak Company's auditorium. The members of the Rochester Chapter found Dr. Guertler's talk very interesting and instructive. We feel sure that all the members who listened to his talk got some good pointers on alloys. *Carl F. Wattel.*

ROCKFORD CHAPTER

The Rockford Chapter opened its 1926-27 season Friday evening, Oct. 1, with a supper at the Nelson Hotel. After the business of the meeting, consisting of a report on the Chicago Convention and reports of the various standing committees, was carried through, the floor was then turned over to Mr. Peck, chairman of the program committee, who introduced G. E. Harcke of the industrial engineering department of the Air Reduction Sales Company.

Mr. Harcke's subject "Oxygen—the Wonder Worker" was illustrated by experiments with liquid oxygen and accompanied by four reels of moving pictures showing the process of obtaining oxygen from the air and the uses of the oxyacetelene flame. Although oxygen as a gas does not lend itself easily to demonstrating because it is colorless and odorless, oxygen as a liquid does things that are readily observable and quite out of the ordinary. The evening was both interesting and instructive and was enjoyed by over sixty members and friends.

R. M. Smith.

SCHENECTADY CHAPTER

The first meeting for the season of 1926-1927 of the Schenectady chapter of the American Society for Steel Treating was held on Thursday afternoon, September 30. Thanks to the courtesy of Mr. Hobarth, it took the form of an inspection trip through the plant of the Hudson Valley Coke and Products Corporation, Troy, N. Y. Here a blast furnace and by-product coke ovens were seen in operation, and an incidental view of wrought iron manufacture was obtained in the Burden Iron Company plant.

Some twenty members attended, going and returning by automobile.

M. F. Sayre.

SYRACUSE CHAPTER

This season's first meeting of the Syracuse Chapter was held on the evening of October 12 at the new building of the Syracuse Chamber of Commerce. Immediately after dinner, Charles Howe, one of the most prominent of Syracuse business men, gave a very entertaining and interesting talk telling us why he loved Syracuse.

The speaker of the evening was to have been Howard J. Stagg, assistant manager at Halcomb Steel Company and he was to give his now famous talk "Why Did It Break?" Unfortunately, Mr. Stagg was called out of town and all of us were much disappointed. However, Mr. Stagg gave the manuscript to Dr. "Bill" Frazer, assistant metallurgist at Halcomb and told him to go ahead and "pinch-hit" for him.

And Bill went ahead and made a pretty good job of it. He showed us pictures of various types of failures such as fatigue failures, grinding cracks, cracks due to improper heating and quenching, and failures due to sharp angles, keyways and stamp marks.

In the discussion which followed the talk, questions were shot at Bill from various angles and only once did he seek refuge behind the answer " 'Cause Stagg says so."

In the business part of the meeting, Chairman Walter Bolton urged the members to support the new evening course in Heat Treatment and Metallography which is being given at Syracuse University in cooperation with the Extension Department of Syracuse University and the Syracuse Chapter of the Steel Treaters. The course is modelled after the one given by the Philadelphia Chapter and intends to give the men in the shops and laboratories some of the fundamental principles on which heat treating operations are based. So far the attendance in the course has been encouraging.

S. P. Peskowitz.

WORCESTER CHAPTER

The second regular monthly meeting of the chapter was held the evening of October 6, at Barratti & Ble's Restaurant.

Supper was served at 6:30 to thirty-eight members and guests.

H. Klauke, our representative to the national convention, talked a few minutes on some of the main features of the convention.

Wilbur Searle called the attention of the chapter to the course in metallurgy offered at the Y. M. C. A. engineering school with the cooperation of the Worcester chapter. The course is given by Victor Hillman, metallurgist of the Crompton & Knowles Loom Works.

The principal speaker of the evening was Stanley P. Rockwell of the S. P. Rockwell Company, Hartford, Conn. Mr. Rockwell's subject was the "Relation of Laboratory to Manufacturing Plant," which he illustrated with lantern slides.

In his talk Mr. Rockwell clearly showed how important the laboratory is to the manufacturing plant, calling attention to many causes of failure of steel that can be explained in the laboratory. For locating the cause of failure, he stated that microscopic examination is far more important than chemical analysis. Also that a high school graduate with proper training could carry on this type of laboratory work.

Mr. Rockwell explained the working of the dilatometer method for heat treating.

At the close of the talk he discussed different methods of procedure in locating cause of failure in answer to questions.

C. G. Johnson.

Items of Interest

W. S. Jones has recently resigned as vice-president and director of the Vanadium-Alloys Steel Company.

B. F. Shepherd has been elected on the board of managers of the Engineers' Club of the Lehigh Valley to serve until May, 1929.

The Ludlum Steel Company announces that Philip L. Coddington has joined their New England Sales Organization.

Columbia Tool Steel Company, Milwaukee Branch, have moved from 137 Sycamore Street to much larger space at 109-111 Reed Street. Larger stocks will now be carried in addition to the installation of new equipment.

Dr. Birger Egeberg, formerly metallurgist with the Halcomb Steel Co., Syracuse, N. Y., has joined the laboratory forces of the International Silver Company, Meriden, Conn.

Frank A. Hall, sales engineer, Dempsey furnace division, W. N. Best Corp., New York, was killed Oct. 19 in an automobile accident at Newark, N. J. He was 35 years of age and a well known figure in the industrial furnace industry. Mr. Hall was a graduate of the University of Pennsylvania. He was connected at different times with the Bristol Co., F. J. Ryan and Co., Republic Flow Meter Co. and for the past year has been with the W. N. Best Corp. He was an ardent booster for the A. S. S. T. and a member of the New York Chapter.

William P. Woodside was recently made president of the American Twist Drill and Tool Company, Detroit.

The Nicholson File Works, Anderson, Ind., has named Durbin Hollingsworth, purchasing agent, to fill the vacany created by the death of William Devitt.

James W. Barr has been made Chicago sales engineer for the New Britain Machine Company of New Britain, Conn. He was formerly connected with the New York office of Manning, Maxwell and Moore.

Allis-Chalmers Manufacturing Company of West Allis, Wis., has appointed L. F. Berry to take charge of its new office at Jackson, Michigan. This is a branch of the company's Detroit office.

It has recently been announced that the Cyclops Steel Company of Pennsylvania has purchased the Cyclops Steel Company, Inc., of New York. The new company will operate the plant at Titusville and conduct the business formerly carried on in the name of the Cyclops Steel Company,

(Continued on Page 34 Adv. Sec.)

Employment Service Bureau

This bureau is for all members of the Society. Want ads will be printed at the following rates: minimum of 30 words $0.50; each additional word $0.02.

This service is also for employers, whether members of the Society or not. Rates for this service are as follows: minimum of 50 words $1.00; each additional word $0.02. Fee must accompany copy.

Address answers care of AMERICAN SOCIETY FOR STEEL TREATING, 4600 Prospect Ave., Cleveland, unless otherwise stated.

POSITIONS WANTED

METALLURGIST-CHEMIST, familiar with testing and heat-treatment of automotive steels, gray iron, malleable iron and steel foundry practice, metallographic research, construction of furnaces, with 9 years' experience and capable executive, wants responsible position. Address 11-5.

FORGE SHOP FOREMAN, practical and competent, desires position. Has had 22 years' experience in all kinds of hammer forging, hardening, heat treating and tool maintenance. At present employed, and reasons for change are very good. Will furnish very good references. Address 11-15.

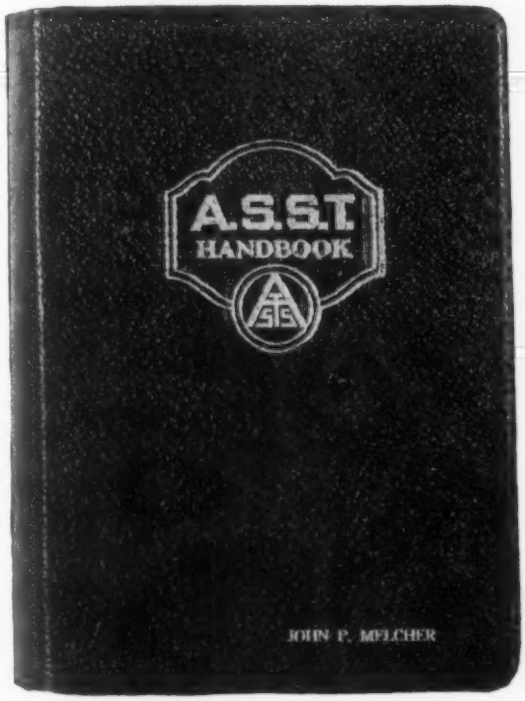
METALLURGIST AND CHEMIST desires change offering future in exchange for 15 years of unusual practical experience in the heat treatment of all types of steel and non-ferrous metals. Familiar with the manufacture of die castings and electro plating processes. Will consider connection in executive or technical sales capacity. Married. Inquiries solicited. Address 11-10.

POSITIONS OPEN

WANTED

SALESMEN. Representation in the leading cities west of Pittsburgh is desired. Applicants must be successful and personally acquainted with the tool steel users in his territory. To such we offer a specialty that has met with instant success. Commission basis. Detroit, Chicago, St. Louis and Milwaukee territory open. Address 11-20.

SALESMEN. Young men with metallurgical training and some experience in this line as tool steel salesmen for an old, established company with headquarters in the Pittsburgh District. Excellent opportunity for advancement. Opening for one or more men with right qualifications. Address 11-25.



ORDER NOW!

DATA SHEET BINDER

The binder shown here is of genuine cowhide, with 1 inch rings, capacity 350 sheets and pocket in cover, especially made for A. S. S. T. Hand Book sheets.

Do not lose your valuable sheets by not having the proper binder.

Members may obtain this binder, with name (in gold) stamped on cover for **$2.00.**

Send order and check to

American Society *for* Steel Treating

Cleveland, Ohio

Inc., of New York and more recently in the name of Charles T. Evans, Lessee.

It is the purpose of this new organization to continue to manufacture the high speed and other tool steels for which the Cyclops plant, instituted by Charles Burgess in 1884, has been famous. The same high standards of quality, service and workmanship, which have always characterized Cyclops Steels, will be assured by the practical steel and business experience of those who are the officers and directors of the new company and by the presence of the same mill personnel who have been trained for many years to this end.

A. F. MacFarland has been appointed as sales representative for the Michigan territory of the Cyclops Steel Company of Titusville, Pa. Mr. MacFarland has for the past six years been connected with the Vanadium Alloys Steel Company as district sales manager and prior to that time he served in the caapcity of metallurgical engineer for this company. He has also been an ardent worker in the interests of the American Society for Steel Treating and in the early days he was editor of the *Journal* of the American Steel Treaters' Society.

The Philadelphia Chapter has recently published its 1926-27 year book. It contains among many other interesting things the tentative program for the year 1926-27 and a detailed description of the evening course in heat treatment and the metallography of steel. This course began on October 4, at Temple University and is designed primarily to be of assistance to those who desire to extend their knowledge while actually engaged in gainful occupations. While the course is scientific, it does not aim to rival the technical colleges in the making of metallurgical engineers. Its aim is rather to foster interest in these subjects and give the student a good understanding of the fundamental principles of them and how these principles may be applied in their work to produce a better product with the expenditure of less time and money. The breadth and scope of the course makes it possible for those who do not have special scholastic preparation but who do possess intelligence, accuracy and industry to benefit greatly by it.

The lecture courses may be taken without the laboratory course. The chapter is awarding scholarships for good work.

The advance class is designed for those who wish to continue after they have finished the full course which consists of the lecture and laboratory courses.

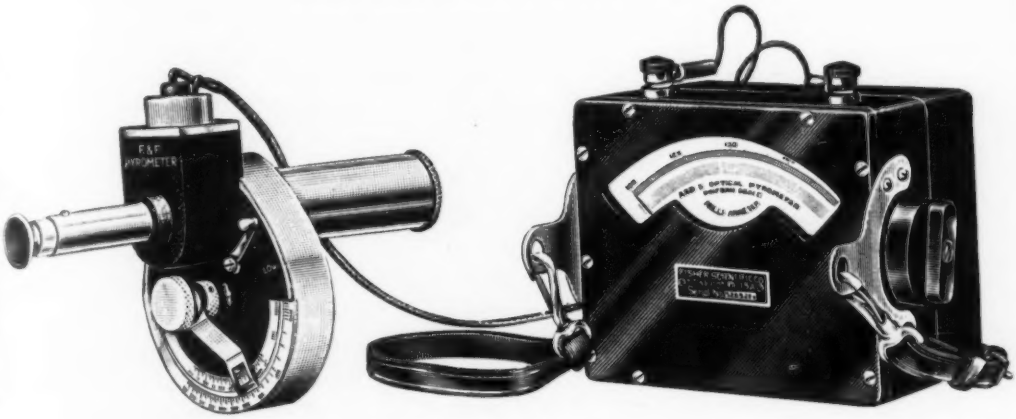
Knowledge is power and special training enables a man to advance more rapidly. His services become more valuable to the world and the world reciprocates in proportion as he is able to give, not only in money but comfort, enjoyment, leisure, respect, etc. Every man should realize that he is on more payrolls than the one in the cashier's office.

It is a well-known psychological fact that groups of individuals work well together and accomplish more than each could accomplish individually. Therefore, the policy adopted by numerous concerns of encouraging their employees to take the course in a group is sound business psychology.

The temperatures of heated objects can be quickly and accurately measured by observing the objects with an

F. & F. OPTICAL PYROMETER

(Another Fisher Product)



One merely observes the heated object through the telescope, rotates the pointer as directed, and reads the temperature on the scale of the instrument.

It is direct reading; no tables required.
Temperatures of molten metals and heated objects can be measured in all stages of operation, such as in the furnace, pouring, rolling, forging, heat treating, etc.

Temperature Range—1600° to 4000° F.

Price $175.00

FISHER SCIENTIFIC COMPANY

LABORATORY SUPPLIES

PITTSBURGH, PA. U.S.A.

FORMERLY SCIENTIFIC MATERIALS CO.

The Carborundum Company of Niagara Falls, N. Y., manufacturers of abrasive products, announces the appointment of W. J. Ulrich as district sales manager of the coated abrasive sales division in the Detroit territory.

Mr. Ulrich will have complete charge of the sales of Carborundum Brand garnet paper and cloth, Carborundum paper and cloth, Aloxite paper and cloth and the Carborundum Brand waterproof finishing papers in the above mentioned district.

Mr. Ulrich has had a wide and varied experience in selling and surfacing coated abrasive products. His appointment became effective October 1.

W. B. Brookfield, for many years superintendent of melting at the Halcomb Steel Company, Syracuse, N. Y., has resigned. His plans for the future are indefinite.

Announcement has been made of the appointment of G. H. Grundy as president of the Poldi Steel Corporation of America, New York City.

R. & G. Import Company, Inc., New York City, has available pamphlets describing the new Baumann-Steinruck impact hardness tester. It contains a description of the construction and working principles of the apparatus; the method of carrying out tests; the selection of intensity of impact and size of the ball; the measurement of ball impressions by microscope and the practical application of measurements obtained. Examples for using the curve tables are given and the checking of the accuracy of calibration of the apparatus and suggestions for its care are set forth.

The Westwater Supply Company, 150 North Third Street, Columbus, Ohio, has been appointed distributor in Columbus and vicinity for Adamant fire brick cement.

In Dayton, Ohio, The Klinger-Dills Co., 129-131 North Jefferson Street, has been appointed distributor for Adamant fire brick cement.

Coan Equipment Company, 236-242 Murray Street, Fort Wayne, Indiana, has been appointed distributor in Fort Wayne and vicinity for Adamant fire brick cement.

The Cleveland Tool & Supply Company, 1427-1437 West Sixth Street, Cleveland, Ohio, will handle and distribute Adamant fire brick cement, effective November 15, 1926.

All four organizations will be glad to cooperate with every user of fire brick interested in prolonging brickwork life and in reducing refractories costs.

Charles M. Schwab, who was made an honorary member of the American Society for Steel Treating at the annual convention in Chicago, September last, has been elected president of the American Society of Mechanical Engineers. He will formally assume his office at the annual meeting in New York in December.

Hevi Duty Electric Company, 690 Kinnickinnic Avenue, Milwaukee, Wisconsin, will occupy new offices, October 18, 1926, at 217 Sycamore Street, Milwaukee.

The New Timken Engineering Journal, a loose-leaf book of 110 pages

Simonds Display at National Steel and Machine Tool Exhibit, Chicago, September 20, 1926

Pioneer Exhibitor Chicago, 1919

September 1, 1926.

Simonds Saw and Steel Co.,
Lockport, N. Y.

Gentlemen:

Because the exhibition for steel and steel treating equipment, supplies and accessories, was born in Chicago in 1919, and because you were one of those who had faith and confidence in that little acorn which has, in a few years, grown to a great oak, and because you had the courage to back that faith with the money to make its birth possible, the Chicago Chapter, the cradle of this exhibition, believes that those pioneer exhibitors should be honored with some mark of distinction, in order that they may be designated among their fellows and to *all* men as the path finders who blazed the trail in making the American Society for Steel Treating the most prominent annual industrial exhibition held in America.

With this thought in mind, we will deliver to your booth on the Municipal Pier, a banner.

We hope you will feel proud to display it at this and future exhibitions held under the auspices of our society. We hope so because we are proud of you.

Sincerely yours,

Executive Committee
Chicago Chapter
AMERICAN SOCIETY FOR STEEL TREATING.

contains technical information relative to the application of Timken Bearings to automotive and industrial machinery. A number of pages are devoted to the explanation of the Timken Bearing as manufactured at the present time. Exclusive features such as the positive alignment of the rolls, one-piece precision cage and special alloy steel are explained.

Typical problems, with the solutions, involving the calculation of various loads and the selection of suitable bearings are given. Tables showing bearing ratings, capacities and dimensions as well as speed capacity curves are included.

Methods of mounting Timken Bearings, shaft and housing design, adjustment of Timken Bearings, closures, cup and cone fitting practices, assembly methods and lubrication are treated in separate chapters.

A full set of dimension sheets accurately drawn to scale, together with formulas and recommendations for the application of Timken Bearings, developed through experience gained in successfully applying more than 150,000,000 bearings comprise another section.

Automotive and industrial engineers will find this journal a valuable reference volume for their technical library. Copies may be had by writing The Timken Roller Bearing Company at Canton, Ohio.

Yale D. Hills has been appointed branch manager of the Seattle branch of The Timken Roller Bearing Service and Sales Company. Mr. Hills was formerly branch manager of the Portland branch. When this branch office was closed he came to the main office at Canton, Ohio, where he has been located until his present appointment.

R. H. Cross who held the position of Seattle branch manager prior to Mr. Hills, has been made assistant to G. C. McMullen, district manager of sales, Industrial Division. Mr. Cross will continue to maintain offices at Seattle.

E. N. Beisheim, formerly of The Bock Bearing Company, Toledo, Ohio, has been appointed assistant to the general manager of The Timken Roller Bearing Service and Sales Company, Canton, Ohio.

S. C. Partridge has been placed in charge of the Buffalo office, Industrial Division, of The Timken Roller Bearing Company. Mr. Partridge takes the place of Lee Warrender, who resigned, to enter another field of work. Mr. Partridge has been in training in the engineering department of The Timken Company and is well versed in the technical problems relative to the application of Timken bearings in industrial machinery.

A celebration of the twenty-fifth anniversary of the Bureau of Standards of the Department of Commerce is just announced. On Saturday, December 4, next, the Bureau will keep Open House and a banquet will be given at which the many friends of the Bureau will meet the staff and reminiscences will be exchanged, the achievements of the quarter century will be reviewed, and the present and future work will be discussed. A group of distinguished guests will attend. The event is of interest to the world of science and as well to the industrial experts who have worked so closely in cooperation with the Bureau and, in turn, made application of its discoveries and developments in perfecting the measured control of processes. The opportunity to inspect the experimental research facilities of the Bureau will be welcomed by its many friends.

November

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